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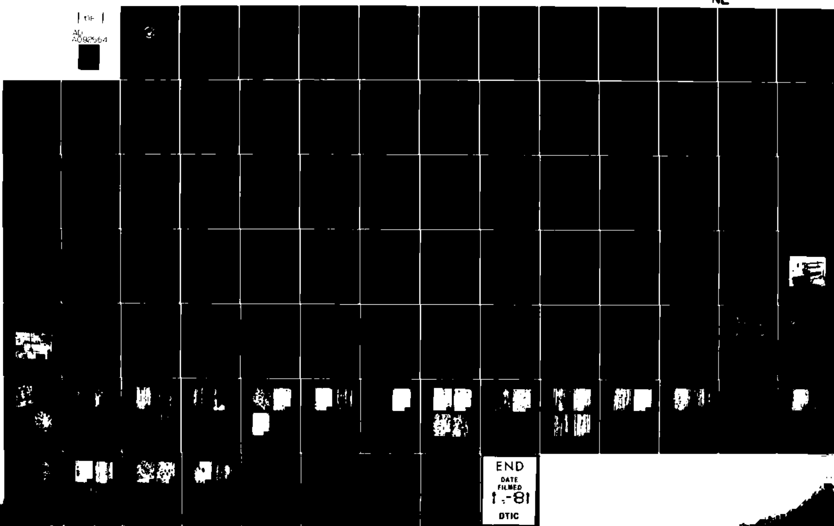

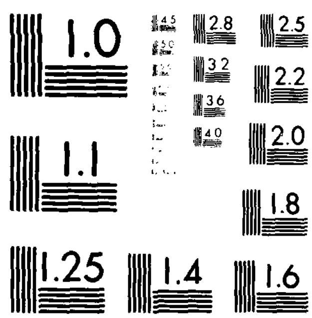

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THESIS

THE INFLUENCE OF ALLOY COMPOSITION AND THERMO-MECHANICAL PROCESSING PROCEDURE ON MICROSTRUCTURE AND MECHANICAL PROPERTIES OF HIGH-MAGNESIUM ALUMINUM-MAGNESIUM ALLOYS

by

Ralph Brian Johnson

June 1980

Thesis Advisor: T. R. McNelley

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The purpose of this research was the development of a thermo-mechanical procedure to process Aluminum-Magnesium alloys and testing effects of alloying additions on these alloys. Magnesium contents of eight and ten weight percent and the alloying effects of copper and manganese were studied. Microstructures and mechanical properties at six warm rolling temperatures located above and below the solvus line of these alloys were examined. Ultimate tensile strengths in excess of 680 MPa (99 KSI) were obtained.

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Microstructural evidence indicated that the precipitation of the "beta" intermetallic phase was one of the most important mechanisms in controlling the strength of the alloy. Furthermore, precipitation is so rapid at higher temperatures that it becomes the strongest force within the microstructure and its presence prevents any possible recrystallization of the alloy. However, when the temperature exceeds the solvus temperature for the alloy, recrystallization does occur with large losses in both yield and ultimate tensile strength.

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The Influence of Alloy Composition and Thermo-mechanical Processing Procedure on Microstructure and Mechanical Properties of High-Magnesium Aluminum-Magnesium Alloys

by

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Submitted in partial fulfillment of the requirements for the degree of

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ABSTRACT

The purpose of this research was the development of a thermomechanical procedure to process Aluminum-Magnesium alloys and testing effects of alloying additions on these alloys. Magnesium contents of eight and ten weight percent and the alloying effects of copper and manganese were studied. Microstructures and mechanical properties at six warm rolling temperatures located above and below the solvus line of these alloys were examined. Ultimate tensile strengths in excess of 680 MPa (99 KSI) were obtained.

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I. INTRODUCTION

The purpose of this thesis was to further investigate the effects of thermo-mechanical processing treatments on the mechanical properties of aluminum magnesium alloys with eight and ten weight percent magnesium and additional alloying elements. Work done by Ness [Ref. 1], Bingay [Ref. 2], Glover [Ref. 3], Grandon [Ref. 4], Speed [Ref. 5] and Chesterman [Ref. 6] at the Naval Postgraduate School has produced strengths exceeding conventional aluminum magnesium alloys. Additionally, Bly, Sherby, and Young's work with hypereutectoid carbon steels [Ref. 7] further indicated that, by thermo-mechanical processing of an alloy, a finer grained micro-structure can be developed that enhances the properties of that alloy. As clearly demonstrated by Grandon [Ref. 4], high strengths are obtainable from high magnesium content aluminum alloys but no standardized processing method was used.

This thesis presents a standardized thermo-mechanical process whereby six high strength Al-Mg alloys were processed and further investigates the effects of additional alloying elements in the eight and ten weight percent magnesium Al-Mg alloys.

II. BACKGROUND

A. PREVIOUS WORK

Ness [Ref. 1] studied an 18% magnesium aluminum alloy and demonstrated that a warm rolling process could produce an alloy with a homogenous microstructure and compressive strengths in excess of 655 MPa (95 KSI). However, the rolling procedure employed was extremely slow and required small reductions per pass.

Bingay [Ref. 2] and Glover [Ref. 3] concurrently followed Ness's work, but instead of manufacturing their own alloys, used a commercial source (Kaiser Aluminum and Chemical Company).

Bingay worked with 15% and 19% magnesium Al-Mg alloys in an attempt to achieve microstructural refinement through isothermal and non-isothermal forging. Neither method achieved this desired microstructural refinement and he recommended that alloys within the range ten to fifteen weight percent magnesium be further investigated.

Glover continued work with the forged 15% to 19% magnesium Al-Mg alloys in an attempt to determine their tensile properties, but the alloys cracked upon rolling. Glover then acquired a Al-11%Mg alloy and successfully rolled it with resulting tensile properties of 4-6% elongation and an ultimate tensile strength of 414 MPa (60 KSI).

Grandon [Ref. 4] studied Al-7%Mg and Al-10%Mg alloys. His forging-rolling process (which did not include the isothermal forging process) produced ultimate tensile strengths in excess of 552 MPa (80 KSI). Grandon, like Bingay, recommended further investigation of alloys with greater than ten weight percent magnesium.

Speed [Ref. 5] followed Grandon's work with a twelve percent aluminum magnesium alloy. However, Speed had severe difficulty in rolling these materials. Speed used the isothermal forging process at the twenty hour point of a twenty-four solid solution heat treatment at 440°C as the first deformation step in his processing.

Chesterman [Ref. 6] performed a recrystallization and precipitation study of the seven, eight, ten, and twelve weight percent magnesium alloys. He found that during warm rolling, these materials do not recrystallize until the rolling temperature was at or above the solvus temperature.

This previous work was used in this thesis research to develop a thermo-mechanical processing sequence that would consistently produce the desired fine microstructures in these aluminum alloys. This process was then used to study the effects of additional alloying elements in the eight and ten weight percent magnesium aluminum alloys.

III. EXPERIMENTAL

A. MATERIAL PROCESSING

1. Melting

Each aluminum ingot was produced by ALCOA using 99.99% pure aluminum base metal and was alloyed to produce the desired compositions using commercially pure copper and magnesium, 75% Mn-Al briquettes, 5% beryllium-aluminum master alloy, and 5% titanium-0.2% boron-aluminum rod. The ingots were produced by the direct chill process using a 127 mm (5 inch) diameter mold. Each ingot was 1016 mm (40 inches) long. Below is a list of the composition of each alloy.

S. No.	Si	Fe	Cu	Mn	Mg	Ti	Be
501299A	0.01	0.03	0.00	0.00	10.2	0.01	0.0002
501300A	0.01	0.03	0.00	0.52	10.2	0.01	0.0002
501301A	0.01	0.03	0.41	0.00	10.3	0.01	0.0002
501302A	0.01	0.03	0.43	0.52	10.4	0.01	0.0002
501303A	0.01	0.03	0.40	0.00	8.14	0.01	0.0002
501304A	0.01	0.03	0.41	0.52	8.22	0.01	0.0002

2. Ingot Homogenization

The solid solution heat treatment of an alloy is performed by heating the alloy into a single phase region in which all other phases in the alloy are dissolved into that single phase. As Figure 1 indicates the eight and ten weight percent Mg-Al alloys are single phase "alpha" at temperatures of 340°C

and 380°C respectively. A temperature of 440°C was selected to ensure complete solid solutionizing of the Al-Mg alloys. However, eight and ten weight percent Al-Mg alloys containing 0.52% Mn require a minimum temperature of 490°C to dissolve the Mn Al_6 intermetallic phase.

3. Isothermal Forging

The isothermal forging of the Al-Mg alloys was done for two reasons: first, to even further reduce the dendritic as cast microstructure through an isothermal compression of three to one; second, to be able to roll the material using the available rolling mill with its limited maximum roll opening of 32 mm (1.25 inches).

4. Warm Rolling Technique

To be able to reduce an Al-Mg alloy from a maximum height of 32 mm (1.25 inches) to a minimum height of 1.75 mm (.07 inches) a standard rolling technique was developed. This technique was used throughout the processing of all the Al-Mg alloys with the exception of the initial Al-8%Mg alloy.

In warm rolling, a temperature drop of 20°C per rolling pass occurs when the billet is removed from the furnace and passed through the cold rolls of the rolling mill [Ref. 4]. To ensure this did not affect the shape (warping) of the billet, a specific rolling sequence was developed. By passing each end of the forged billet through the rolling mill and then by turning

it over and repeating the same pattern, warpage of the sheet was almost negligible.

B. TENSILE TESTING

Tensile blanks were sheared from the rolled alloy into specimen blanks measuring 127 mm (5 inches) by 19 mm (.75 inches). These blanks were then machined into tensile specimens with a gauge length of 37 mm (1.75 inches). Tensile tests were performed using a Model TT-D Instron Floor Model Testing Machine utilizing a crosshead speed of 1.27 mm/min (.05 inch/min).

C. METALLOGRAPHY

All sample mounting for metallographic examination was done using room temperature epoxy resin compounds. Any cutting of specimens was done by wet cutting methods. Polishing of the specimens required wet sanding with 240,400 and 600 grit paper with further polishing using 1.0 micron and 0.05 micron Al_2O_3 slurries. Final polishing was performed using the automatic polisher (vibromet) with magomet slurry. The etchant used was Barker's reagent. All photomicrographs were taken on the Zeiss optical microscope.

IV. RESULTS

A. STANDARDIZED THERMO-MECHANICAL PROCESSING SCHEDULE

The processing of the Al-8%Mg alloy was the initial investigation of this thesis. From Speed's [Ref. 5] prior work, a solid solution homogenizing time of twenty-four hours at 440°C was used. After twenty hours, the 32 by 32 by 96 mm (1.25 by 1.25 by 3.75 inch) cast billets were isothermally forged with a three to one reduction in height. These forged billets were then placed back into the furnace completing the twenty-four hour solution treatment time. After twenty-four hours at 440°C, the forged billets were oil quenched.

Six of the forged billets were then given a 50% reduction by cold rolling and six were left as-forged. These twelve Al-8%Mg alloy billets were warm rolled at temperatures of 200°C, 250°C and 300°C using two different rolling schedules, one with thirty minutes between rolling passes and another using five minutes between rolling passes.

Further experimentation with the Al-8%Mg alloy found that the solution treatment time of twenty-four hours at 440°C could be reduced to ten hours at 440°C with the isothermal forging occurring at the nine hour point of the solution temperature treatment.

Before the warm rolling began on the forged billets, each billet received a ten minute preheat time at the warm rolling

temperature. Furthermore, by placing a thermocouple on the forged billets it was determined that a two minute reheat time between rolling passes was adequate.

Subsequent thermo-mechanical processing of the 8% and 10% magnesium Al-Mg alloys was done using the ten hours at 440^{o}C solid solution homogenizing treatment with an isothermal forge at the nine hour point and a final oil quench. The forged billets were then given a ten minute preheat at warm rolling temperature and a rolling schedule of two minutes in furnace between rolling passes. All forged billets were rolled to a final thickness between 1.75 to 2.25 mm (.07 to .09 inches) with a final water quench.

B. MECHANICAL PROPERTIES

The Al-Mg alloys displayed a certain range of interesting mechanical properties ranging from very high yield strengths of 612 MPa (88 KSI) to various changes in yield and ultimate tensile strengths due to alloying additions.

Al-8%Mg was the first Al-Mg alloy mechanically tested with results appearing in Figure 4. The Al-8%Mg alloy was used for a comparison of cold work versus warm worked mechanical properties, including the study of longer interpass times during warm rolling (Section IV A). Table I shows these results and, clearly, nothing was gained by the longer interpass times.

However, a higher yield and ultimate tensile strengths were obtained by using prior cold work, but cold rolling was extremely difficult and therefore eliminated.

Al-8.14%Mg-0.42%Cu was the first alloy mechanically tested using the standard thermo-mechanical process (Section IV A). A 34 MPa (5 KSI) increase was noted in both yield and ultimate tensile strengths over the Al-8%Mg alloy. Table II and Figure 5 show the mechanical properties of Al-8.14%Mg-0.42%Cu alloy and reveal a consistent trend of reduced strength and increased ductility as warm rolling temperature is increased.

The addition of 0.52%Mn to this Al-8.14%Mg-0.42%Cu alloy results in a large increase in tensile strength. The manganese addition increased the ultimate and yield tensile strengths to 596.3 MPa (86.5 KSI) and 540 MPa (78.4 KSI) respectively while retaining a ductility of 6.0%. This result is amazing in that no other Al-Mg alloy previously processed by warm rolling ever achieved such a good combination of strength and ductility. Table III and Figure 6 show the mechanical properties of the Al-8.22%Mg-0.42%Cu-0.52%Mn alloy. Figure 11 reveals that the yield strength after warm rolling at 200°C is increased 116.5 MPa (16.9 KSI) over the Al-8.14%Mg-0.42%Cu alloy and a 153 MPa (22.2 KSI) increase in yield strength over the Al-8%Mg alloy.

The ten percent Al-Mg alloy study began with the binary alloy, Al-10.2%Mg. Table IV and Figure 7 present the tensile

properties of this alloy. The addition of two percent more magnesium has increased the tensile properties of the Al-8%Mg alloy and matched the Al-8.14%Mg-0.40%Cu alloy, (Figures 11, 12, 13). Al-10.2%Mg exhibited an excellent ductility of 25% at the 440°C warm rolling temperature, but its yield tensile strength dropped to 270 MPa (39.1 KSI). This alloy did not exhibit the increased ductility with increasing warm rolling temperature, but ductility remained constant at about 14% until the warm rolling temperature reached 440°C.

The addition of 0.41%Cu increased the yield and ultimate tensile strengths over the Al-10.2%Mg alloy by about 41.4 MPa (6 KSI), Figure 11. The copper alloying addition again increased the alloy's strength with only a slight loss in ductility. Figure 8 presents the mechanical property data of the Al-10.3%Mg-0.41%Cu alloy. The Al-10.3%Mg-0.41%Cu alloy continually lost both yield and ultimate strengths as warm rolling temperature was increased. However, the ductility gradually decreased with increasing warm rolling temperature until at 400°C it increased dramatically.

Al-10.2%Mg-0.52%Mn alloy clearly exhibited the strengthening effect of the manganese. At the warm rolling temperature of 200°C its strength was over 131 MPa (19 KSI) higher than the Al-10.2%Mg alloy, Figure 11 . However, ductility was reduced from about 12% (Al-10.2%Mg) to 6%. The important fact is, however, that this alloy is stronger than any commercially

produced aluminum alloy, heat treatable or not. The effect of warm rolling temperature on strength and ductility is shown in Figure 9 and Table VI for the Al-10.2%Mg-0.52%Mn alloy, a continual decrease in strength was evident as warm rolling temperature was increased.

The last alloy processed was the Al-10.4%Mg-0.52%Mn-0.43%Cu alloy. Alloys containing magnesium as the primary alloying agent are referred to as a 5000 series alloy; 5083-H343 which is the strongest 5000 series alloy has a yield strength of 262 MPa (38 KSI), Figure 14. The Al-10.4%Mg-0.52%Mn-0.43%Cu alloy has a yield strength of 614 MPa (88.8 KSI) over 200% higher than the strongest Al-Mg alloy. Furthermore, its strength is exceeding the best commercially heat treatable alloy 7075-T6. These results indicate that a great increase in strength can be obtained by careful combination of alloying and thermo-mechanical processing. The Al-10.4%Mg-0.52%Mn-0.43%Cu alloy is by far the strongest alloy ever produced by warm rolling. A complete list of mechanical properties is compiled in Figure 10 and Table 7. The major concern with this alloy is its limited ductility, 6%.

A puzzling result is the fact that the Al-Mg alloys containing manganese decrease in ductility when warm rolled above 340°C even though the strength is continually decreasing. These results are unique to the Al-10%Mg-0.52%Mn alloy with and without the copper addition. Furthermore, physical

evidence on the tested tensile specimens indicated a tendency to break at the gauge length scribe marks, indicating possible notch sensitivity at these warm-rolling temperatures.

V. DISCUSSION OF RESULTS

A. Al-8.14%Mg-0.4%Cu

The as-cast microstructure in Figure 15-1-a shows precipitation both at the grain boundaries and within the grains. From the phase diagrams for the Al-Mg binary system [Ref. 8] and the Al-Cu-Mg ternary system [Ref. 9], the precipitate is mostly of "beta" (Mg_5Al_8), with the possibility of some $CuMg_4Al_6$. After the standard solution heat-treatment (10 hrs at 440°C) the precipitate is almost completely dissolved. Figure 17 shows the effect of cooling rate from the solution treatment temperature. The oil quenched specimen has more precipitate, principally at the grain boundaries, than the water quenched specimen. This "beta" precipitate has precipitated on cooling from the solution treatment temperature.

Figure 18-1 shows the microstructures of the alloy after warm-working. At the temperatures 200°-380°C a banded microstructure can be seen, with different amounts of precipitate in each band. This is caused by segregation at the ingot stage into the magnesium-rich and magnesium-lean areas. The precipitate volume fraction will be higher in the magnesium-rich areas than elsewhere.

The precipitate is fine at low temperatures of warm rolling, and coarsens as the temperature rises. This is to be expected, as low precipitation temperatures should favor nucleation

of new precipitate particles, while higher precipitation temperatures favor growth of existing precipitate particles.

Since the fine precipitates will tend to be close together, the flow stress of the specimens rolled at low temperatures should be greater than those rolled at high temperatures, since the flow stress will be proportional to 1/d, (where d = inter-particle spacing).

The specimen rolled at 420°C showed a recrystallized grain structure, and had less strength and a greater ductility that the specimens rolled at lower temperatures. This series of micrographs confirmed Chesterman's observation [Ref. 6] that recrystallization was prevented from occurring by the presence of closely spaced precipitate particles. When the solvus temperature was exceeded, the specimen recrystallized on rolling.

B. Al-8.22%Mg-0.52%Mn-0.41%Cu

The as-cast structure of this alloy Figure 15-1-b was similar to that of the previous alloy (Figure 15-1-a). In this case, solution treatment at 440°C dissolved some of the precipitate but the rest failed to dissolve. It was necessary to go to 490°C to completely solution treat the alloy. By examining the Al-Mn binary diagram [Ref. 10] and the Al-Mg-Mn ternary diagram [Ref. 11], it can be seen that the "beta" (Mg_5Al_8) dissolves at 440°C, while another intermetallic $MnAl_6$ is present up to 490°C.

Figures 20-1 and 20-2 show the microstructures of the warm rolled specimens. As for the previous alloy, the "beta" precipitate particles are fine for the lower rolling temperatures and coarsen as the temperature rises. At 420°C, the precipitate volume fraction is reduced, and the widely spaced fine particles have failed to prevent recrystallization of the alloy.

The size and volume fraction of precipitate can be seen more clearly in Figures 21-1 and 21-2, at the higher magnification. The highest strength obtained was at 200°C where the precipitates were finest, being hardly resolvable in the light microscope.

C. Al-10.2%Mg

The as-cast structure was similar to previous alloys. The warm rolled alloys showed a higher amount of "beta" (Mg_5Al_8) precipitation that the 8%Mg alloys, as would be expected from the increased magnesium content. Again, as the temperature of rolling increased, the coarseness of the "beta" precipitates increased, until, at 440°C, the alloy recrystallized (Figure 22-2-f). The precipitate appearing at the grain boundaries and within the grain at 440°C probably formed during the oil quench from the rolling temperature.

D. Al-10.3%Mg-0.41%Cu

Compared with the Al-10.2%Mg alloy, the addition of 0.41%Cu increased the volume fraction of precipitate. By analogy with

the Al-8.14%Mg-0.4%Cu alloy most of the precipitate will be "beta" (Mg_5Al_8) with a greater possibility of some $CuMg_4Al_6$. This alloy had a reduced volume fraction of precipitate at 440°C, but the volume fraction was higher than that of the Al-8.14%Mg-0.4%Cu and the Al-10.2%Mg. These latter two alloys recrystallized, while the Al-10.3%Mg-0.41%Cu did not.

The effect of the copper seems to be to increase the volume fraction of precipitate and also to slow down the dissolution of the precipitate at the solution treatment temperature.

E. Al-10.2%Mg-0.52%Mn

It was necessary to modify the solution treatment for this alloy, as was done for the Al-8.22%Mg-0.52%Mn-0.41%Cu, so as to dissolve the $MnAl_6$ particles at 490°C.

Figures 24-1 and 24-2 show the microstructure at 100X, while Figures 25-1 and 25-2 are taken at 1000X. Compared with previous alloys, there is little visible within-grain precipitation. The white precipitate visible at a warm rolling temperature of 200°C probably comes from the rolling out of "beta" formed during oil quenching of the billet from 490°C. Between 250°C and 400°C, the precipitates get steadily coarser, as in all the other alloys of the series. At 440°C, the volume fraction of precipitate is very much reduced. The fine precipitate particles, probably $MnAl_6$, have been so widely spaced that partial recrystallization was noticed.

While the alloy was being rolled at 440°C, it was noticed that a black liquid was oozing out of the forged billet during rolling. This indicated that there could be incipient melting occurring during working. Mondolfo [Ref. 11] reports a possible ternary eutectic: liq $\longrightarrow$ Al + Mg_5Al_8(beta) + $(MgMn)_3Al_{10}$ at 437°C and 33%Mg, 0.5%Mn with balance of Al. With the high magnesium content (10%) and manganese present, it may be possible to have the liquid enriched sufficiently during the initial chill cast to solidify at the grain boundaries as the ternary eutectic. For alloys rolled at low temperatures, the eutectic may well remain in the interdendritic areas that it occupied on solidification, while for alloys rolled at higher temperatures it could be rolled out (or flow out at a temperature of 440°C) and form a thin layer over much of the grain boundary area of the alloy.

F. Al-10.4%Mg-0.52%Mn-0.43%Cu

This quaternary alloy was heat treated at 490°C to solution any $MnAl_6$ present. The microstructure follows the same pattern as the other alloys in the series, with fine precipitation at low temperatures and coarser precipitation at high temperatures. Compared with the Al-10.2%Mg-0.43%Cu alloy, the volume fraction of precipiate was greater, especially at higher temperature. The higher volume fraction, and more closely spaced precipitate

prevented any recrystallization of the alloy at 440°C. This effect of the copper was the same as that between the Al-10.2%Mg and Al-10.3%Mg-0.41%Cu alloys noted previously.

Again, as in the Al-10.2%Mg-0.52%Mn, oozing of black liquid occurred during rolling at 440°C, raising the same suspicions about a ternary eutectic.

VI. EFFECTS OF MICROSTRUCTURE ON MECHANICAL PROPERTIES

A. EFFECTS OF WARM ROLLING TEMPERATURE

Chesterman [Ref. 6] found through a precipitation study of these alloys that hardness increased as time at temperature increased. This is precipitation hardening as explained in Ref. 12. The alloy increased in hardness until a specified time at temperature at which it overaged. The overaging occurs because the fine particulate intermetallic "beta" phase grows to such a size that it has increased the inter-particle spacing. However, only slight increases in hardness were documentated in comparison to the hardness increases gained by warm rolling.

Warm rolling of the alloys at low temperatures will increase the dislocation density of the alloy. This additional pinning of the microstructure by the precipitate particles may have formed dislocation cells and inhibited recrystallization of the grains. But, when warm rolling temperatures increased to a point where recrystallization could occur (above the solvus), recrystallization reduced the dislocation density and hence a large decrease was noted in both yield and ultimate tensile strengths.

By virtue of the physical mechanism of warm rolling both precipitation hardening and increased dislocation density occur. Therefore a large increase in hardening of the alloy is noted along with increased tensile strengths.

B. EFFECTS OF ALLOYING ADDITIONS

Magnesium was the primary alloying addition within this study. As shown in comparison between the 8% and 10% magnesium alloys an increase in strength accompanies an increase in magnesium content. The volume fraction of the "beta" precipitate (Mg_5Al_8) accordingly increased with increasing magnesium content. From $\tau \approx \frac{2Gb}{d}$ where τ = shear stress for plastic deformation, G = shear modulus, b = Burgers' vector and d = inter-particle spacing it can be seen that an increased volume fraction of precipitate produces a higher shear stress due to the decreased inter-particle distance assuming the same particle size. Again this was confirmed by the increased strengths of the Al-10%Mg alloys.

Adding copper to the alloys increased the strength over straight Al-Mg alloys, since the presence of copper increased the volume fraction of "beta." Copper additionally acted as an inhibitor towards dissolution of the "beta" phase as experienced in the Al-10.3%Mg-0.41%Cu alloy at 440°C. With this additional inhibiting of the "beta" phase, the volume fraction of precipitate was increased at high warm rolling temperatures which led to higher strengths.

Manganese additions greatly increased the volume fraction of precipitates by the formation of two precipitates, $MnAl_6$ and Mg_5Al_8. These two precipitates increased the volume fraction of the precipitates which again led to very high increases in strengths.

C. LOW DUCTILITY OCCURRENCES

Very low ductilities occurred within the Al-10%Mg alloys when additionally alloyed with manganese above 300°C. Possible answers to this low ductility problem are as follows:

1. Formation of the ternary eutectic would cause the materials to be brittle due to the incipient melting of this phase during the solution treatment. The ternary would then flow throughout the microstructure and form a thin layer across the microstructure and embrittle it, if rolled at a high temperature. The probability of this phase occurring was substantiated even more as the Al-8%Mg alloy with manganese had good ductility, but, when used with Al-10%Mg alloys ductility is almost nil. One method to check this idea would be to use Auger Spectroscopy to determine the grain boundary composition.
2. The presence of coarse $MnAl_6$ could possibly cause the low ductility but, how such small amounts of coarse $MnAl_6$ could cause brittleness when large amounts of coarse "beta" did not in the Al-10.3%Mg-0.41%Cu is difficult to explain.
3. Finally if the presence of "beta" precipitate allied with $MnAl_6$ or the ternary eutectic at the grain boundaries caused brittleness, then subsequent experimental testing should eliminate the oil quench. Using a water quench should eliminate the grain boundary precipitation of "beta" due to slow cooling rates from solution treatment temperatures.

VII. CONCLUSIONS

The purpose of this thesis was to develop a thermo-mechanical processing procedure and then use it to process a series of Al-Mg alloys. Both these goals have been accomplished, but additional questions have been raised about the Al-Mg alloy system.

Yield strengths in excess of 200% stronger than the strongest 5000 series commercial alloy were obtained using a warm rolling procedure only. But before warm rolling could be utilized, investigation into the proper solution heat treatment was carried out with isothermal forging occurring as part of the process. After these steps were thoroughly investigated, a complete metallurgical investigation into the microstructure was done to ensure that no "beta" intermetallic was present before the warm rolling sequence commenced. Further experimentation was done on the rolling procedure and a standard reduction per pass was dependent upon the hardness of the alloy. Additionally, a two minute reheat interpass time was found to be sufficient between passes on the rolling mill.

Controlled precipitation of the fine "beta" intermetallic was found to be the single most important factor when warm rolling these alloys to obtain their best strength and ductility. The most suitable rolling temperatures for these properties have been found to be between 200°C and 340°C.

Recrystallization of the Al-Mg alloys was possible but only at temperatures above the solvus. This led to increased grain growth and high losses in both ultimate and yield strengths.

VIII. RECOMMENDATIONS

It is highly recommended that this research be continued because of the impressive results these alloys display in both ultimate and yield strengths over commercially available alloys of today.

Further recommendations are as follows:

1. A study of the fatigue characteristics of these alloys is essential to ensure that their properties are as good as the other 5000 series alloys.
2. Further experimentation into the corrosion properties of these alloys to ensure the high magnesium content or the additional alloying elements are not causing detrimental effects.
3. If manganese is to continue to serve as an alloying agent, a complete investigation into the possible formation of the ternary eutectic must be made. This would be possible using Auger Spectroscopy.
4. A study of the effects of quenching rates must be undertaken to ensure that no intermetallic is forming at the grain boundaries due to the slowness of an oil quench. This may be eliminated by using a water quench from solution treatment temperatures instead of the standard oil quench.

5. A study of the kinetics of precipitation, both when being warm rolled and in the absence of rolling should be performed. This needs to be done using a standardized thermo-mechanical treatment, and the substructure (precipitate-dislocation) examined by transmission electron microscopy.

TABLE I

Table of Thermo-mechanical Processes
Mechanical Testing Results
Al-8%Mg Alloy

Table Key

OQ: Oil Quench

WQ: Water Quench

WW: Warm Rolled

CW: Cold Rolled

ST: Solution Treated

XXX/XX: Temperature °C/Hours

XX/XX/XXX: Number of Passes/Thousandths of an inch per pass/ Temperature °C

Process	Sample Number	Ultimate Tensile Strength MPa	Ultimate Tensile Strength KSI	0.2% Offset Yield Strength MPa	0.2% Offset Yield Strength KSI	Elonga-tion,%	Hardness R_B
ST 440/24 OQ CW 11/40/20 CW 6/20/20 WW 8/40/200 WW 1/30/200 WQ	5001-1 5001-2	500 497	72.5 72.1	435 438	63.1 63.5	6.25 6.28	82.5
ST 440/24 OQ CW 8/40/20 CW 6/20/20 WW 7/40/250 WW 1/25/250 WQ	5002-1 5002-2	447 458	64.8 66.4	359 365	52.0 52.9	9.38 9.38	80.5
ST 440/24 OQ CW 10/40/20 CW 6/20/20 WW 10/40/300 WW 1/20/300 WQ	5003-1 5003-2	439 440	63.7 63.8	360 345	52.2 50.1	9.38 10.42	77.0

Process	Sample Number	Ultimate Tensile Strength MPa	Ultimate Tensile Strength KSI	0.2% Offset Yield Strength MPa	0.2% Offset Yield Strength KSI	Elongation, %	Hardness R_B
ST 440/24 OQ	5004-1	493	71.5	438	63.6	4.17	
CW 8/40/20	5004-2	493	71.5	436	63.3	4.17	
CW 6/20/20							
WW 4/80/200							
WW 2/40/200							
WW 3/10/200							
WQ							
ST 440/24 OQ	5005-1	448	65.0	388	56.3	7.30	
CW 10/40/20	5005-2	438	63.6	364	52.8	4.17	
CW 6/20/20							
WW 4/60/250							
WW 4/40/250							
WW 1/23/250							
WQ							
ST 440/24 OQ	5006-1	419	60.7	335	48.6	8.33	
CW 8/40/20	5006-2	399	57.9	317	46.0	10.42	
CW 6/20/20							
WW 4/60/300							
WW 4/40/300							
WW 1/35/300							
WQ							
ST 440/24 OQ	5007-1	445	64.5	384	55.7	8.33	
WW 12/60/200	5007-1	461	66.9	391	56.7	6.25	78.75
WW 1/40/200							
WW 1/30/200							
WQ							
ST 440/24 OQ	5008-1	430	62.4	364	52.8	8.34	
WW 21/40/250	5008-2	433	62.8	364	52.8	8.34	80.0
WW 1/20/250							
WQ							
ST 440/24 OQ	5009-1	424	61.5	347	50.4	9.38	
WW 9/60/300	5009-2	403	58.5	328	47.6	12.50	77.0
WW 7/40/300							
WW 1/20/300							
WQ							
ST 440/24 OQ	5010-1	443	64.2	378	54.8	8.33	
WW 2/80/200	5010-2	438	63.6	377	54.7	8.33	
WW 9/60/200							
WW 2/40/200							
WQ							

Process	Sample Number	Ultimate Tensile Strength MPa	KSI	0.2% Offset Yield Strength MPa	KSI	Elongation, %	Hardness R_B
ST 440/24 OQ	5011-1	430	62.3	384	55.7	4.17	
WW 5/80/250	5011-2	440	64.4	374	54.2	8.33	
WW 5/60/250							
WW 4/40/250							
WW 1/23/250							
WQ							
ST 440/24 OQ	5012-1	350	50.8	265	38.4	9.3	
WW 7/80/300	5012-2	393	57.0	319	46.3	10.42	
WW 4/60/300							
WW 2/40/300							
WW 1/25/300							
WQ							

TABLE II

Table of Thermo-mechanical Processes
Mechanical Testing Results
Al-8.14%Mg-0.4%Cu Alloy

Table Key

OQ: Oil Quench

WQ: Water Quench

WW: Warm Rolled

ST: Solution Treated

XXX/XX: Temperature ^{o}C/Hours

XX/XX/XXX: Number of Passes/Thousandths of an inch per pass/ Temperature ^{o}C

Process	Sample Number	Ultimate Tensile Strength MPa	KSI	0.2% Offset Yield Strength MPa	KSI	Elongation, %	Hardness R_B
ST 440/10 OQ	0301-1	490	71.1	419	60.7	11.67	
WW 20/40/200	0301-2	487	70.6	430	62.4	8.70	79.3
WW 1/30/200	0301-3	496	71.9	424	61.5	11.33	
WW 1/25/200							
WQ							
ST 440/10 OQ	0302-1	451	65.4	357	51.8	10.40	
WW 16/40/250	0302-2	459	66.6	382	55.4	10.40	73.5
WW 17/20/250	0302-3	454	65.9	360	52.2	10.00	
WW 1/30/250							
WQ							
ST 440/10 OQ	0303-1	467	67.8	362	52.5	10.40	
WW 18/40/300	0303-2	497	72.1	388	56.3	10.40	73.0
WW 15/20/300	0303-3	473	68.6	398	57.7	10.00	
WW 1/15/300							
WQ							
ST 440/10 OQ	0304-1	430	62.3	329	47.7	15.33	
WW 14/40/340	0304-2	432	62.6	348	50.5	13.33	70.3
WW 15/20/340	0304-3	433	62.8	344	49.9	15.33	
WW 1/30/340							
WQ							

Process	Sample Number	Ultimate Tensile Strength MPa	KSI	0.2% Offset Yield Strength MPa	KSI	Elongation, %	Hardness R_B
ST 440/10 OQ	0305-1	425	61.6	313	45.4	17.70	
WW 16/40/380	0305-2	421	61.0	321	46.6	14.58	65.0
WW 11/20/380	0305-3	413	59.9	289	41.9	14.58	
WW 1/30/380							
WQ							
ST 440/10 OQ	0306-1	379	54.9	251	36.4	21.88	
WW 16/40/420	0306-2	374	54.2	248	35.9	21.88	60.0
WW 16/20/420	0306-3	375	54.4	244	35.4	20.83	
WW 1/15/420							
WQ							

TABLE III

Table of Thermo-mechanical Processes
Mechanical Testing Results
Al-8.22%Mg-0.52%Mn-0.41%Cu Alloy

Table Key

OQ: Oil Quench

WQ: Water Quench

WW: Warm Rolled

ST: Solution Treated

XXX/XX: Temperature °C/Hours

XX/XX/XXX: Number of passes/Thousandths of an inch per pass/ Temperature °C

Process	Sample Number	Ultimate Tensile Strength MPa	KSI	0.2% Offset Yield Strength MPa	KSI	Elonga-tión, %	Hardness R_B
ST 440/10 OQ	0401-1	595	86.3	547	79.4	5.33	
ST 490/3 OQ	0401-2	596	86.5	535	77.6	6.25	86.0
WW 23/40/200	0401-3	598	86.8	538	78.1	6.25	
WW 5/20/200							
WW 1/25/200							
WQ							
ST 440/10 OQ	0402-1	539	78.2	456	66.1	6.9	
ST 490/3 OQ	0402-2	545	79.1	462	67.0	8.33	82.0
WW 16/40/250	0402-3	547	79.3	472	68.4	8.33	
WW 17/20/250							
WW 1/30/250							
WQ							
ST 440/10 OQ	0403-1	510	74.0	417	60.5	8.33	
ST 490/3 OQ	0403-2	530	76.8	428	62.1	8.33	76.7
WW 18/40/300	0403-3	523	75.8	423	61.3	8.33	
WW 15/20/300							
WW 1/15/300							
WQ							
ST 440/10 OQ	0404-1	528	76.6	445	64.5	8.33	
ST 490/3 OQ	0404-2	527	76.4	441	64.0	8.33	78.5
WW 16/40/340	0404-3	519	75.3	467	67.7	7.3	
WW 15/20/340							
WW 1/30/340							
WQ							

Process	Sample Number	Ultimate Tensile Strength MPa	Ultimate Tensile Strength KSI	0.2% Offset Yield Strength MPa	0.2% Offset Yield Strength KSI	Elonga-tion, %	Hardness R_B
ST 440/10 OQ	0405-1	467	67.7	350	50.7	11.46	
ST 490/3 OQ	0405-2	461	66.8	359	52.1	11.46	75.0
WW 21/40/380	0405-3	473	68.6	374	54.3	11.48	
WW 11/20/380							
WW 1/30/380							
WQ							
ST 440/10 OQ	0406-1	424	61.5	263	38.2	18.75	
ST 490/3	0406-2	425	61.6	278	40.3	15.63	65.5
WW 16/40/420	0406-3	426	61.8	299	43.3	17.71	
WW 16/20/420							
WW 1/15/420							
WQ							

TABLE IV

Table of Thermo-mechanical Processes
Mechanical Testing Results
Al-10.2%Mg Alloy

Table Key

OQ: Oil Quench

WQ: Water Quench

WW: Warm Rolled

ST: Solution Treated

XXX/XX: Temperature °C/Hours

XX/XX/XXX: Number of passes/Thousandths of an inch per pass/ Temperature °C

Process	Sample Number	Ultimate Tensile Strength MPa	Ultimate Tensile Strength KSI	0.2% Offset Yield Strength MPa	0.2% Offset Yield Strength KSI	Elongation, %	Hardness R_B
ST 440/10 OQ	9901-1	490	71.1	427	62.0	10.67	
WW 23/40/200	9901-2	493	71.5	419	60.8	11.64	77.0
WW 5/20/200	9901-3	495	71.8	420	60.9	12.50	
WW 1/25/200							
WQ							
ST 440/10 OQ	9902-1	453	65.7	330	47.9	16.67	
WW 16/40/250	9902-2	447	64.9	354	51.3	16.67	74.5
WW 17/20/250	9902-3	450	65.2	343	49.7	15.63	
WW 1/30/250							
WQ							
ST 440/10 OQ	9903-1	471	68.3	373	54.0	12.5	
WW 18/40/300	9903-2	474	56.7	391	56.7	9.4	76.0
WW 15/20/300	9903-3	472	68.5	350	50.7	10.42	
WW 1/15/300							
WQ							
ST 440/10 OQ	9904-1	447	64.9	336	48.7	13.54	
WW 22/40/340	9904-2	453	65.7	337	48.9	16.67	72.5
WW 3/20/340	9904-3	453	65.7	336	48.8	16.67	
WW 1/35/340							
WQ							

Process	Sample Number	Ultimate Tensile Strength MPa	KSI	0.2% Offset Yield Strength MPa	KSI	Elongation, %	Hardness R_B
ST 440/40 OQ	9905-1	451	65.4	333	48.3	11.45	
WW 20/40/400	9905-2	447	64.8	330	47.9	11.45	72.0
WW 7/20/400	9905-3	460	66.7	285	41.4	12.5	
WW 1/32/400							
OQ							
ST 440/10 OQ	9906-1	414	60.0	257	37.3	25.0	
WW 20/40/440	9906-2	414	60.0	261	37.8	25.0	64.0
WW 7/20/440	9906-3	419	60.7	292	42.3	24.0	
WW 1/25/440							
WQ							

TABLE V

Table of Thermo-mechanical Processes
Mechanical Testing Results
Al-10.3%Mg-0.41%Cu Alloy

Table Key

OQ: Oil Quench

WQ: Water Quench

WW: Warm Rolled

ST: Solution Treated

XXX/XX: Temperature °C/Hours

XX/XX/XXX: Number of passes/Thousandths of an inch per pass/ Temperature °C

Process	Sample Number	Ultimate Tensile Strength MPa	KSI	0.2% Offset Yield Strength MPa	KSI	Elongation, %	Hardness R_B
ST 440/10 OQ	0101-1	533	77.3	465	67.4	10.0	
WW 20/40/200	0101-2	541	78.4	464	67.3	10.42	82.0
WW 1/25/200	0101-3	531	77.0	465	67.4	10.42	
WW 1/30/200							
WQ							
ST 440/10 OQ	0102-1	519	75.3	432	62.7	10.42	
WW 16/40/250	0102-2	519	75.3	434	62.9	10.42	81.0
WW 17/20/250	0102-3	531	77.0	434	62.9	10.42	
WW 1/30/250							
WQ							
ST 440/10 OQ	0103-1	488	70.8	409	59.3	8.33	
WW 18/40/300	0103-2	512	74.2	411	59.6	8.33	78.0
WW 15/20/300	0103-3	513	74.4	423	61.4	6.25	
WW 1/15/300							
WQ							
ST 440/10 OQ	0104-1	476	69.1	377	54.7	6.25	
WW 25/40/340	0104-2	478	69.4	379	54.9	6.25	77.0
WW 3/20/340	0104-3	482	69.9	370	53.7	6.25	
WW 1/35/340							
WQ							

Process	Sample Number	Ultimate Tensile Strength MPa	KSI	0.2% Offset Yield Strength MPa	KSI	Elongation, %	Hardness R_B
ST 440/10 OQ	0105-1	460	66.7	321	46.6	17.71	
WW 20/40/400	0105-2	466	67.6	347	50.3	10.42	75.0
WW 7/20/400	0105-3	472	68.4	351	50.9	9.40	
WW 1/32/400							
WQ							
ST 440/10 OQ	0106-1	444	64.4	272	39.5	17.71	
WW 20/40/440	0106-2	445	64.5	292	42.4	17.71	67.0
WW 7/20/440	0106-3	432	62.7	267	38.7	23.46	
WW 1/25/440							
WQ							

TABLE VI

Table of Thermo-mechanical Processes
Mechanical Testing Results
Al-10.2%Mg-0.52%Mn Alloy

Table Key

OQ: Oil Quench

WQ: Water Quench

WW: Warm Rolled

ST: Solution Treated

XXX/XX: Temperature °C/Hours

XX/XX/XXX: Number of passes/Thousandths of an inch per pass/ Temperature °C

Process	Sample Number	Ultimate Tensile Strength MPa	KSI	0.2% Offset Yield Strength MPa	KSI	Elongation, %	Hardness R_B
ST 440/10 OQ	0001-1	634	92.0	556	80.6	6.25	
ST 490/3 OQ	0001-2	630	91.4	557	80.7	7.3	89.0
WW 19/40/200	0001-3	621	90.1	549	79.6	5.2	
WW 4/20/200							
WW 1/15/200							
WQ							
ST 440/10 OQ	0002-1	603	87.5	525	76.2	5.2	
ST 490/3 OQ	0002-2	608	88.2	521	75.5	8.3	95.0
WW 19/40/250	0002-3	603	87.5	466	67.6	6.3	
WW 10/20/250							
WW 1/30/250							
WQ							
ST 440/10 OQ	0003-1	589	85.4	481	69.8	7.3	
ST 490/3 OQ	0003-2	579	84.0	470	68.2	7.3	91.0
WW 21/40/300	0003-3	574	83.3	458	66.4	7.3	
WW 3/20/300							
WQ							
ST 440/10 OQ	0004-1	571	82.8	442	64.1	8.3	
ST 490/3 OQ	0004-2	567	82.3	468	67.9	6.3	91.0
WW 18/40/340	0004-3	554	80.4	439	63.7	7.3	
WW 3/20/340							
WW 1/30/340							
WQ							

Process	Sample Number	Ultimate Tensile Strength MPa	KSI	0.2% Offset Yield Strength MPa	KSI	Elongation, %	Hardness R_B
ST 440/10 OQ	0005-1	483	70.1	403	58.4	3.1	
ST 490/3 OQ	0005-2	547	79.4	450	65.2	2.1	87.5
WW 16/40/400	0005-3	526	76.3	446	64.7	2.1	
WW 4/20/400							
WW 1/30/400							
WW 1/10/400							
WQ							
ST 440/10 OQ	0006-1	520	75.4	427	62.0	4.2	
ST 490/3 OQ	0006-2	523	75.9	436	63.2	4.2	85.0
WW 22/40/440	0006-3	512	74.2	390	56.6	16.7	
WW 2/20/440							
WW 1/16/440							
WQ							

TABLE VII

Table of Thermo-mechanical Processes
Mechanical Testing Results
Al-10.4%Mg-0.52%Mn-0.43%Cu Alloy

Table Key

OQ: Oil Quench

WQ: Water Quench

WW: Warm Rolled

ST: Solution Treated

XXX/XX: Temperature °C/Hours

XX/XX/XXX: Number of passes/Thousandths of an inch per pass/ Temperature °C

Process	Sample Number	Ultimate Tensile Strength MPa	Ultimate Tensile Strength KSI	0.2% Offset Yield Strength MPa	0.2% Offset Yield Strength KSI	Elongation, %	Hardness R_B
ST 440/10 OQ	0201-1	687	99.6	614	89.1	4.2	
ST 490/3 OQ	0201-2	674	97.8	610	88.5	4.2	96.0
WW 39/20/200	0201-3	685	99.3	612	88.8	4.2	
WW 20/10/200							
WQ							
ST 440/10 OQ	0202-1	656	95.2	573	83.1	7.3	
ST 490/3 OQ	0202-2	643	93.3	535	77.6	5.2	96.5
WW 40/20/250	0202-3	643	93.3	565	84.0	5.2	
WW 1/30/250							
WQ							
ST 440/10 OQ	0203-1	621	90.1	522	75.7	5.2	
ST 490/3 OQ	0203-2	601	87.0	492	71.3	5.2	88.0
WW 41/20/300	0203-3	603	87.4	498	72.2	6.3	
WW 1/30/300							
WQ							
ST 440/10 OQ	0204-1	549	79.6	483	70.1	2.1	
ST 490/3 OQ	0204-2	572	82.9	476	69.1	3.1	83.5
WW 49/20/340	0204-3	535	77.6	478	69.4	2.1	
WW 1/30/340							
WQ							

Process	Sample Number	Ultimate Tensile Strength MPa	KSI	0.2% Offset Yield Strength MPa	KSI	Elongation, %	Hardness R_B
ST 440/10 OQ	0205-1	545	79.1	467	67.8	2.1	
ST 490/3 OQ	0205-2	539	78.2	470	68.1	2.1	81.0
WW 45/20/400	0205-3	551	79.9	461	66.9	2.1	
WW 1/30/400							
WQ							
ST 440/10 OQ	0206-1	481	69.8	447	64.9	1.0	
ST 490/3 OQ	0206-2	516	74.9	445	64.6	2.1	78.0
WW 39/20/440	0206-3	514	74.6	410	59.4	2.1	
WW 1/30/440							
WQ							

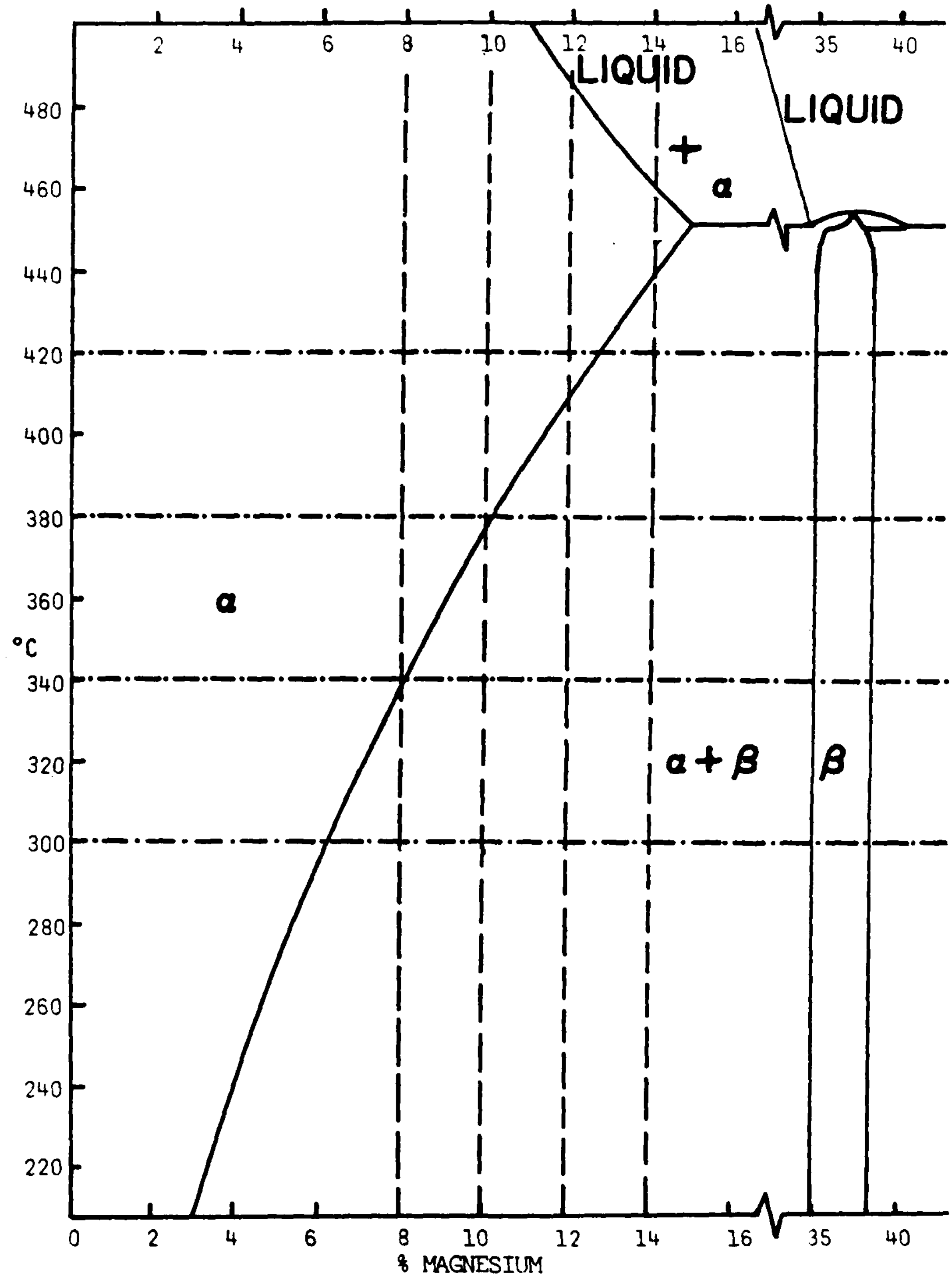


Fig. 1 A partial Aluminum-Magnesium phase diagram. Compositions are indicated with dashed lines, temperatures are indicated by dot-dashed lines.

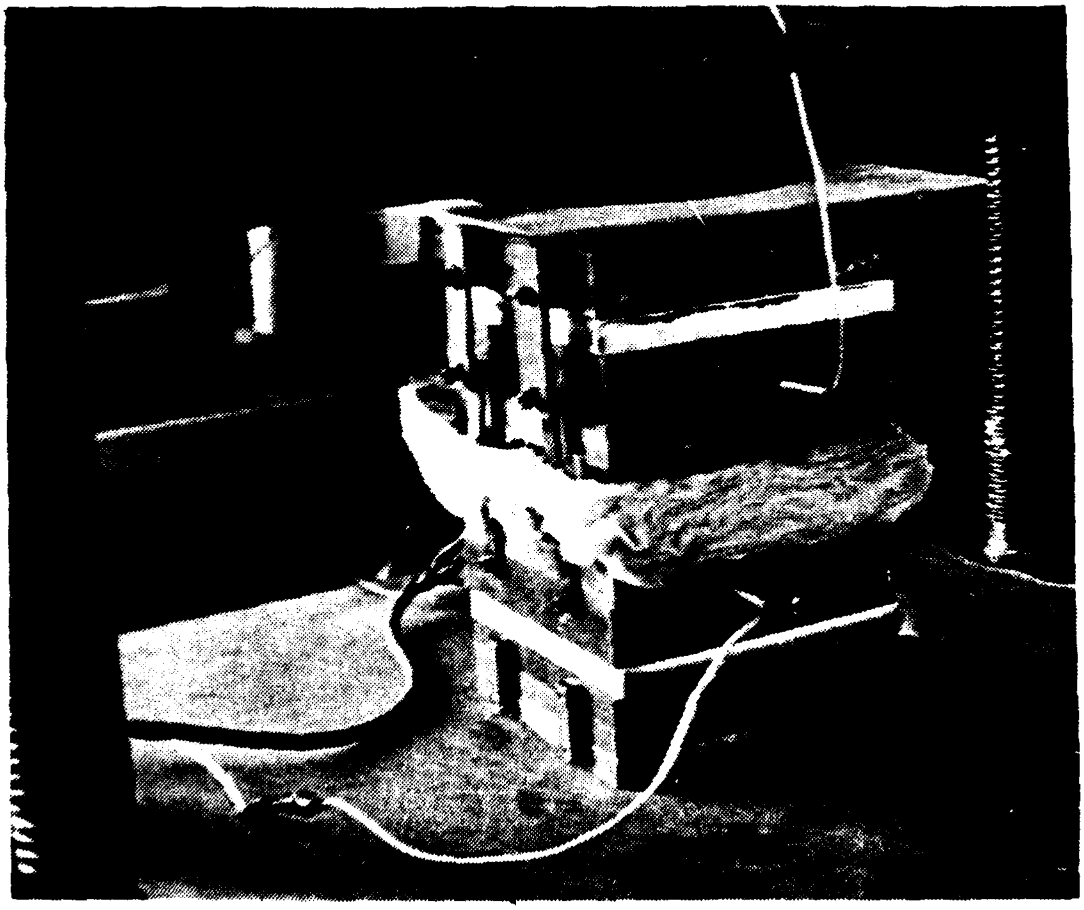


Fig. 2 Photograph of Forging Assembly

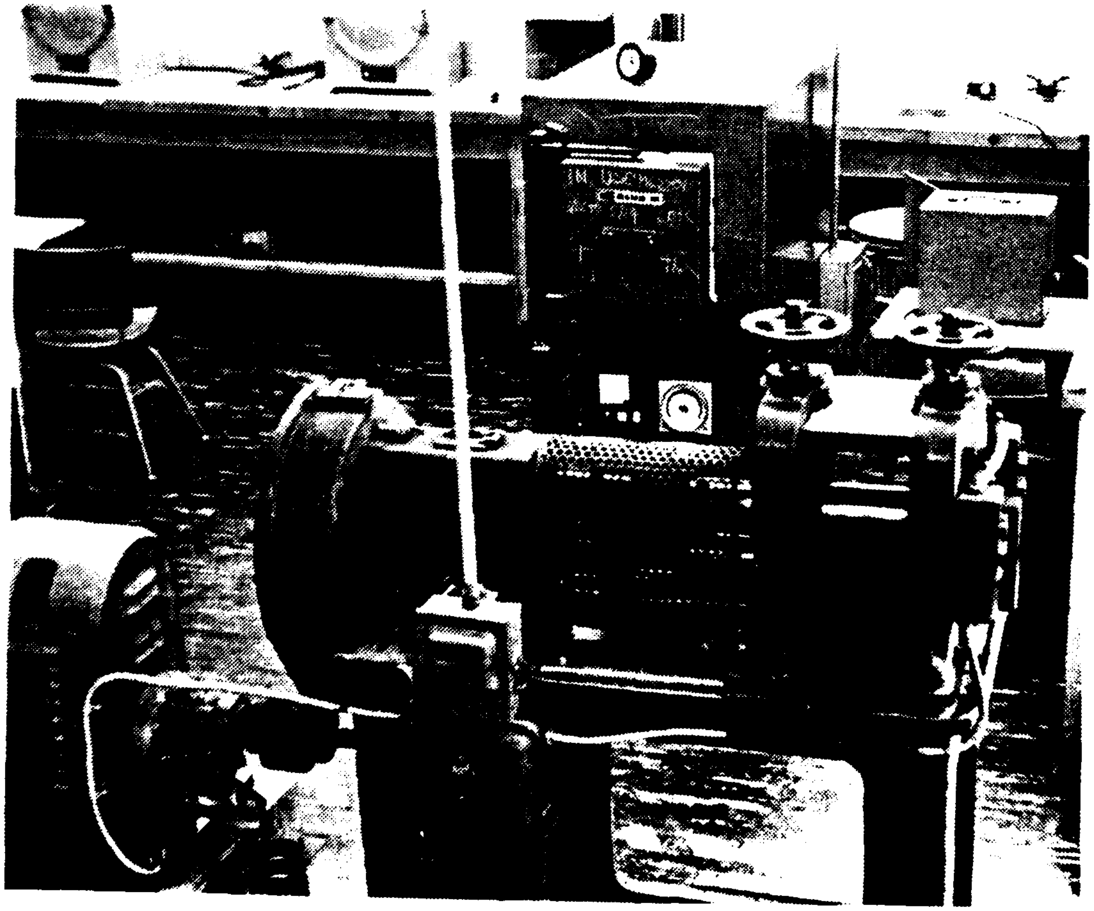


Fig. 3 Photograph of furnace and rolling mill used in the processing of these alloys.

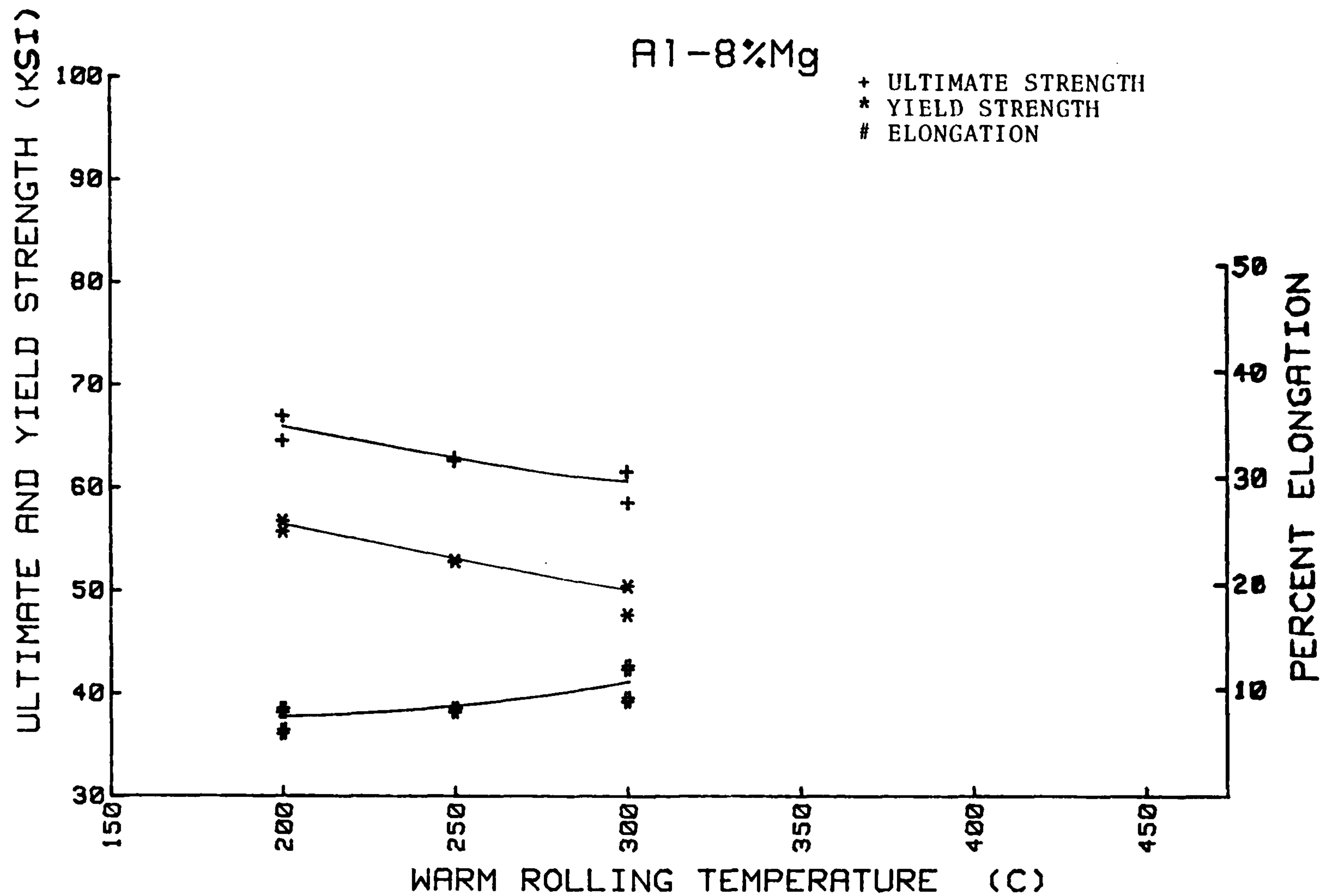


Fig. 4 Al-8%Mg alloy graph of ultimate and yield tensile strengths, elongation, versus warm rolling temperature.

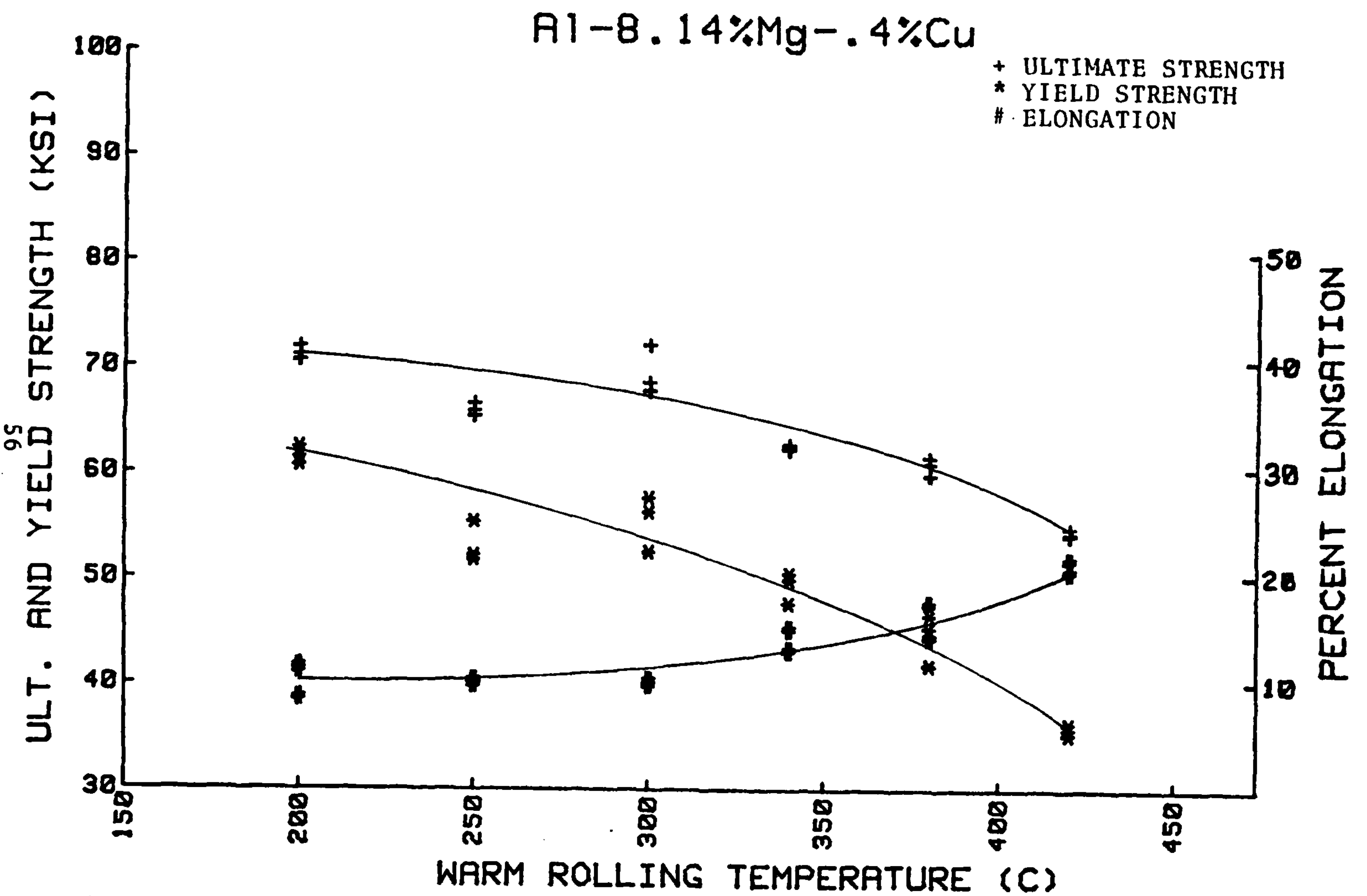


Fig. 5 Al-8.14%Mg-0.40%Cu alloy graph of ultimate and tensile yield strengths, elongation, versus warm rolling temperature.

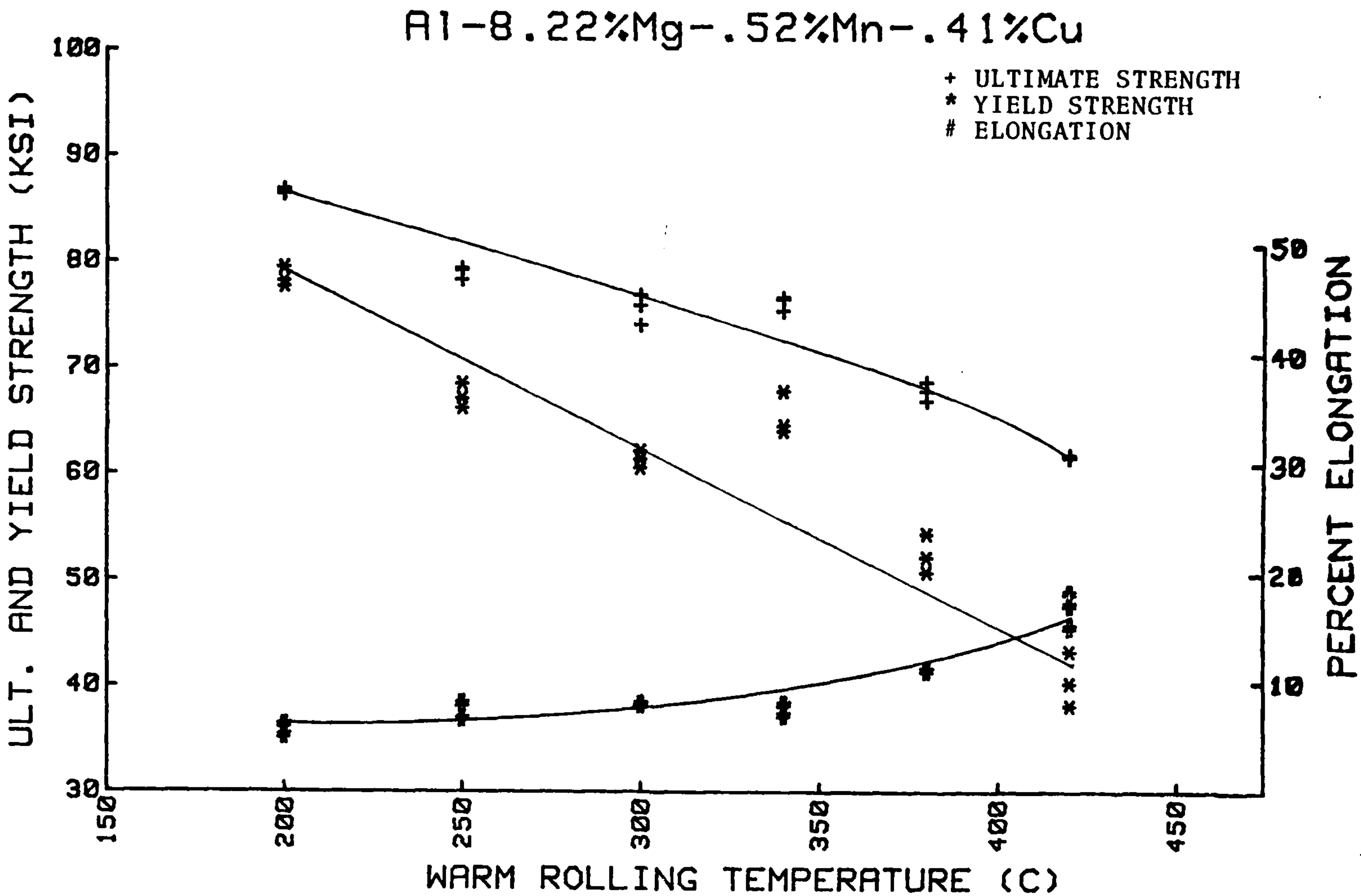


Fig. 6 Al-8.22%Mg-0.52%Mn-0.43%Cu alloy graph of ultimate and yield tensile strengths, elongation, versus warm rolling temperature.

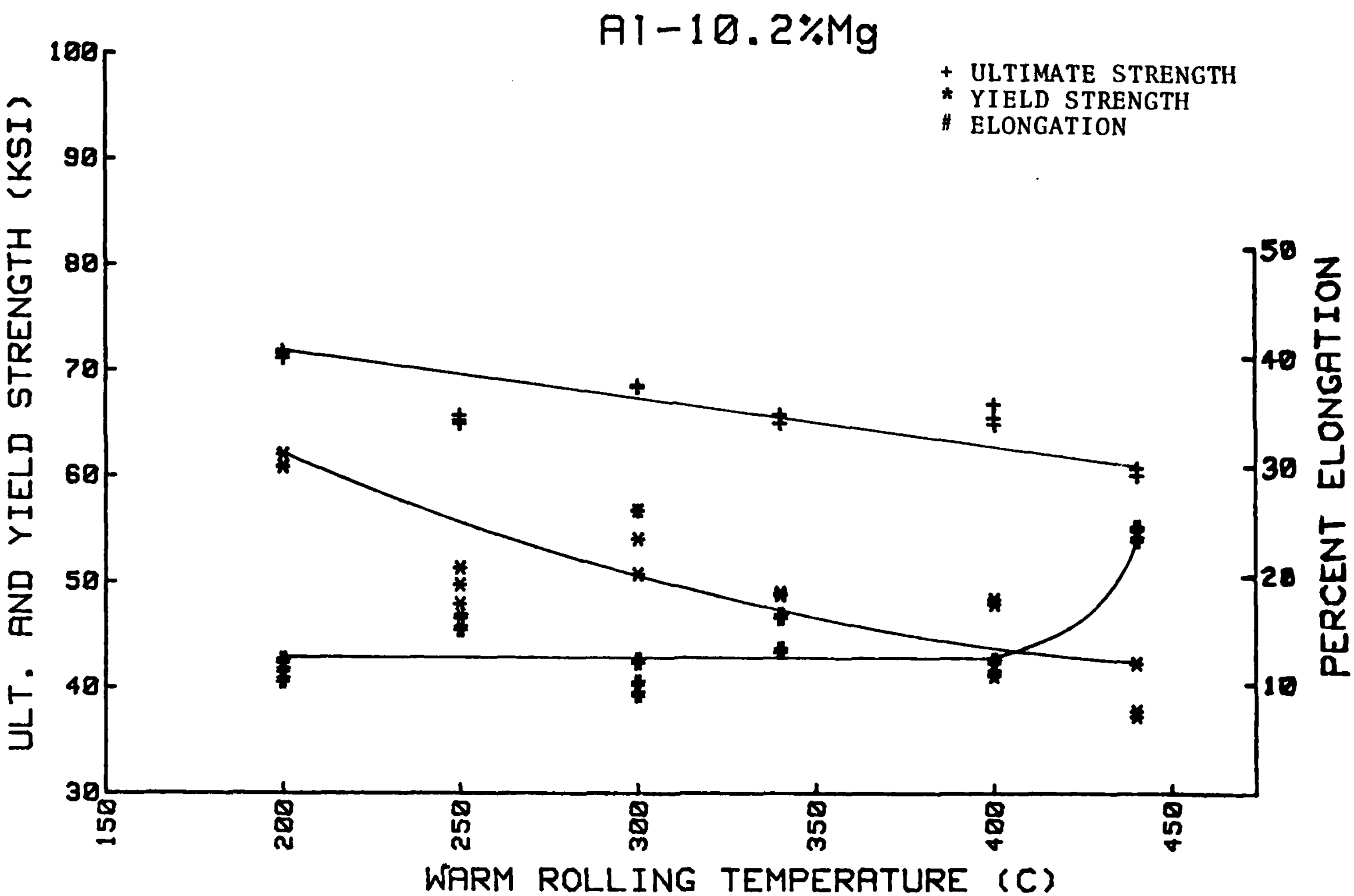


Fig. 7 Al-10.2%Mg alloy graph of ultimate and yield tensile strengths, elongation, versus warm rolling temperature.

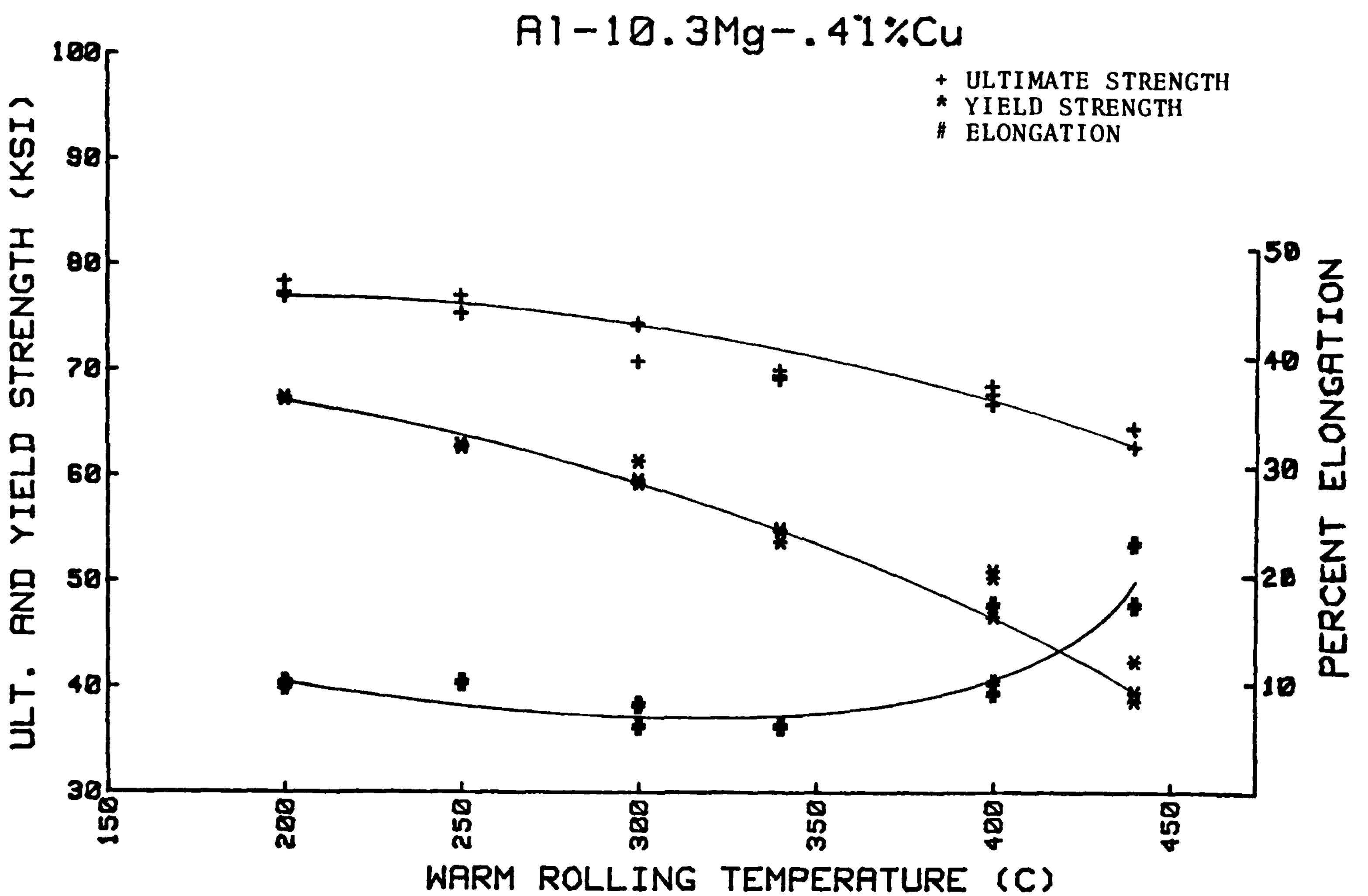


Fig. 8 Al-10.3%Mg-0.41%Cu alloy graph of ultimate and yield tensile strengths, elongation, versus warm rolling temperature.

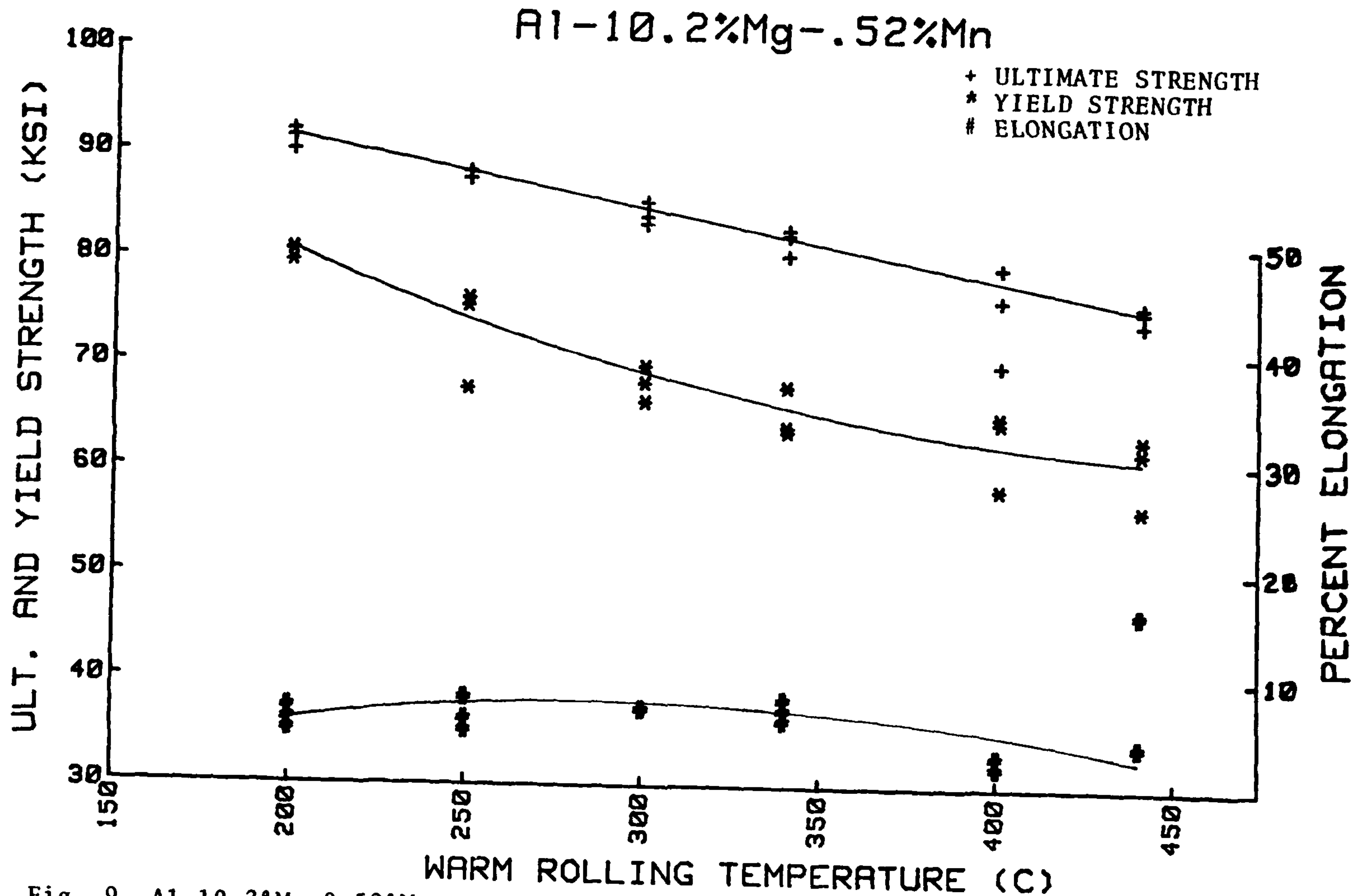


Fig. 9 Al-10.2%Mg-0.52%Mn alloy graph of ultimate and yield tensile strengths, elongation, versus warm rolling temperature.

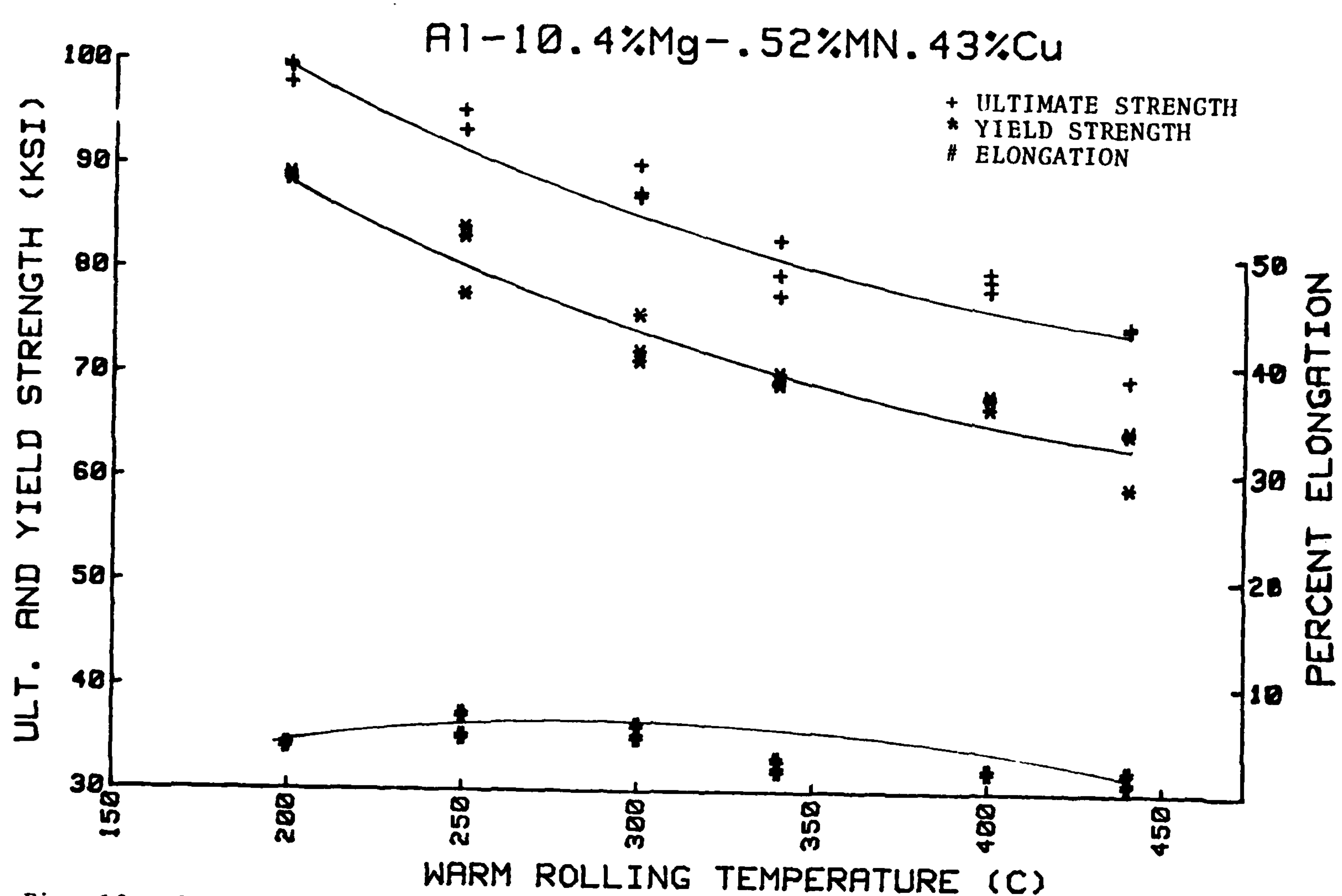


Fig. 10 Al-10.4%Mg-0.52%Mn-0.41%Cu alloy graph of ultimate and yield strengths, elongation, versus warm rolling temperature.

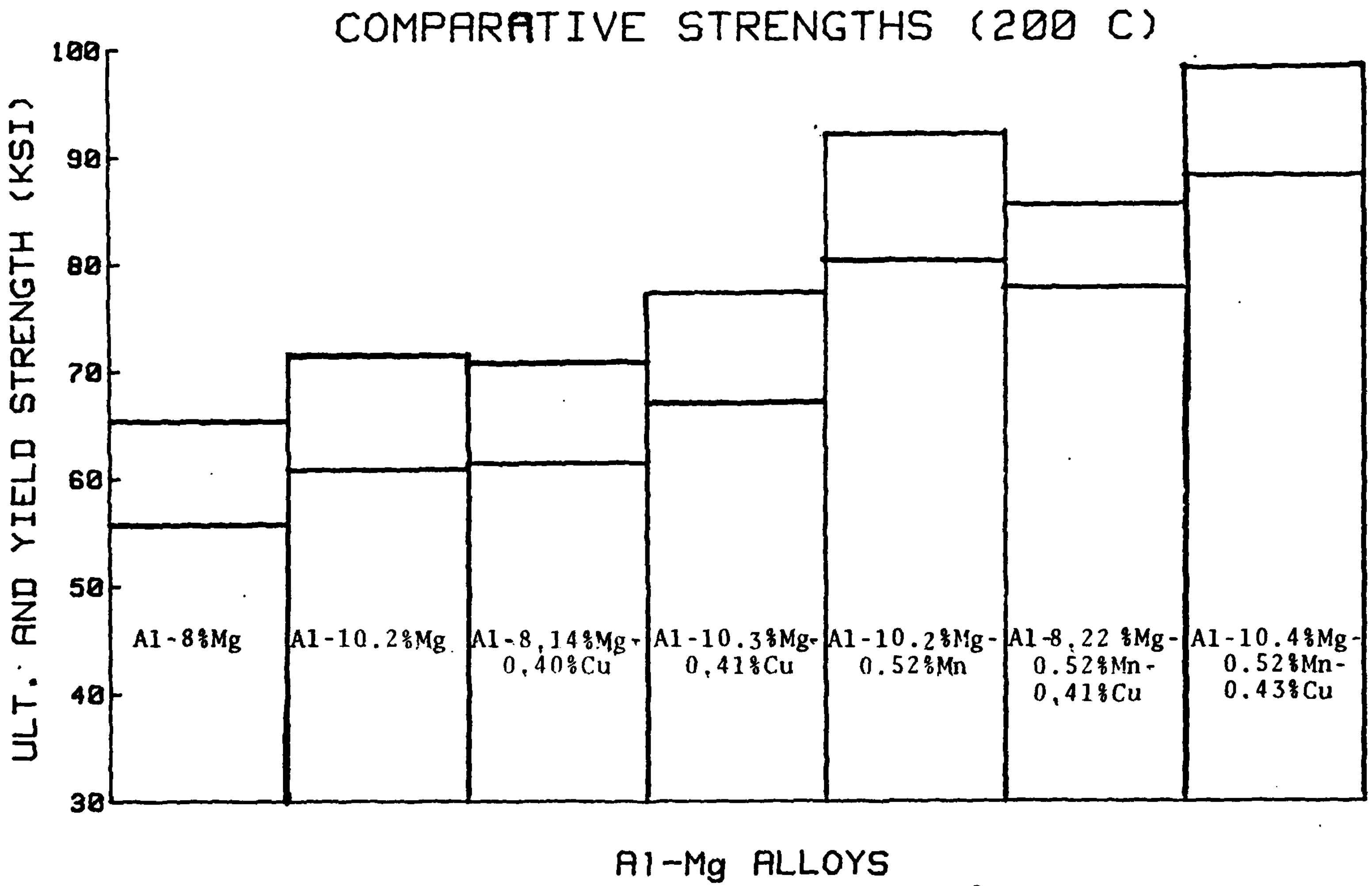


Fig. 11 Graph of comparative strengths for all alloys at 200°C warm rolling temperature.

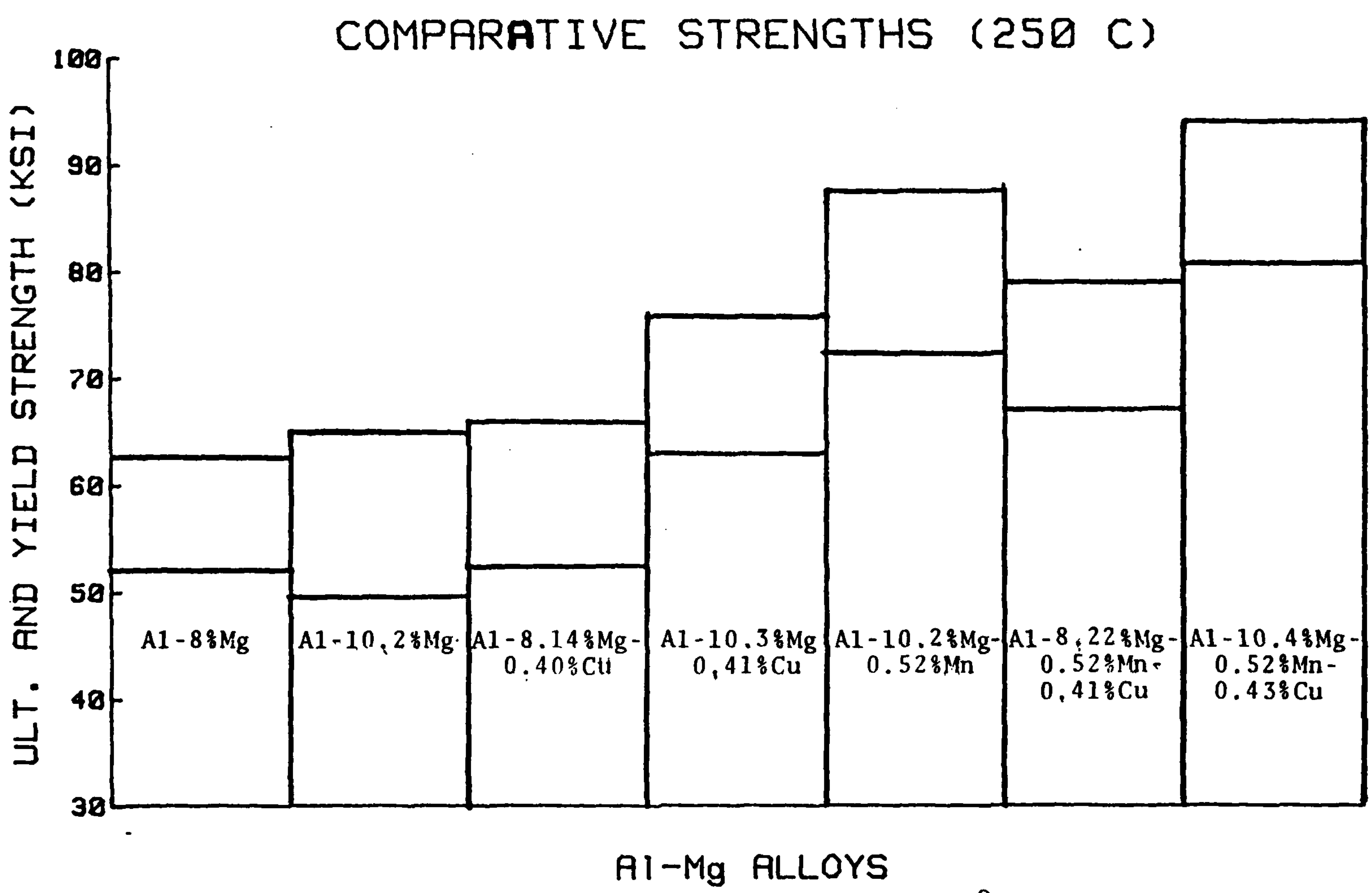


Fig. 12 Graph of comparative strengths for all alloys at 250°C warm rolling temperature.

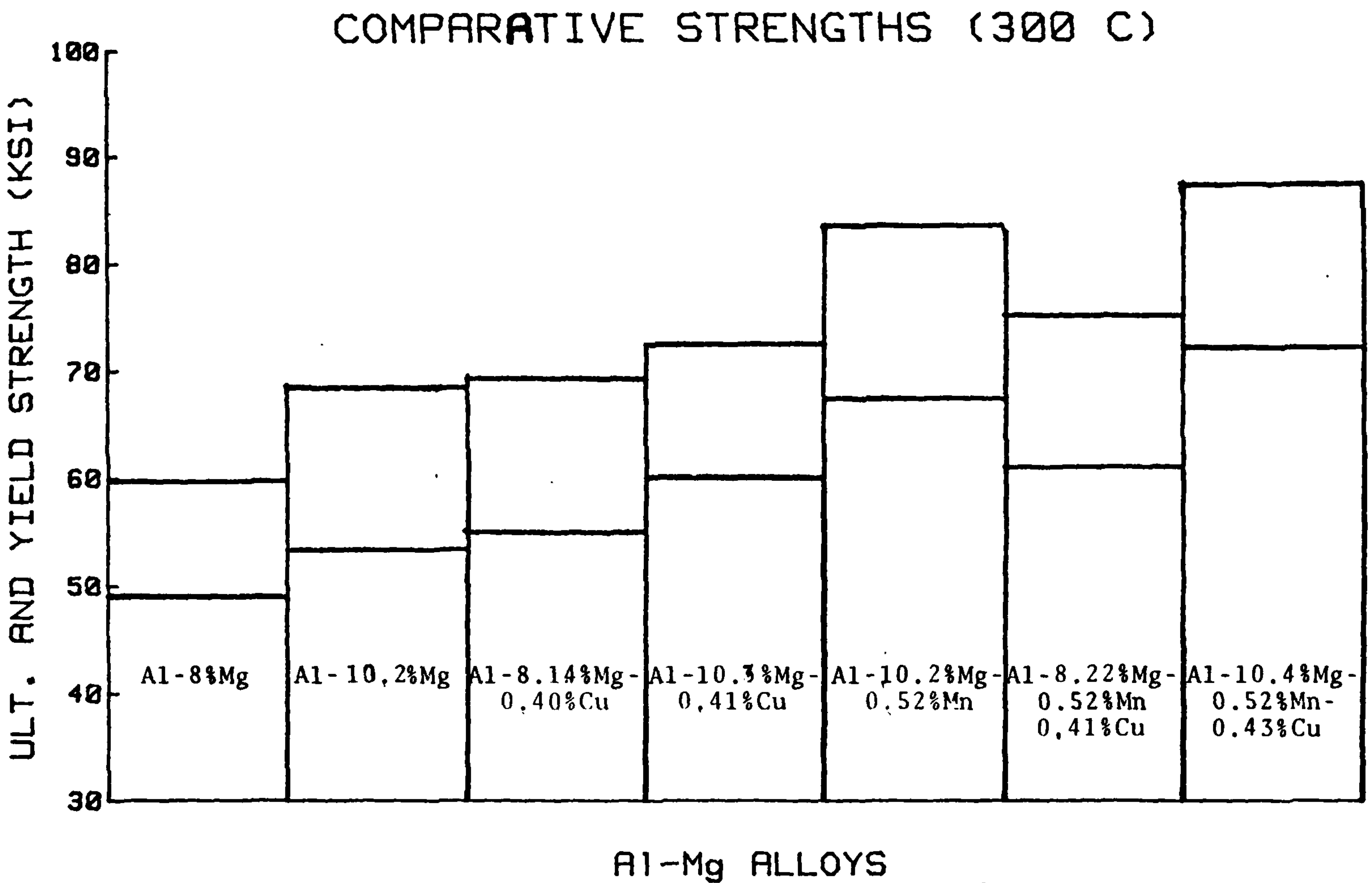


Fig. 13 Graph of comparative strengths for all alloys at 300°C warm rolling temperature.

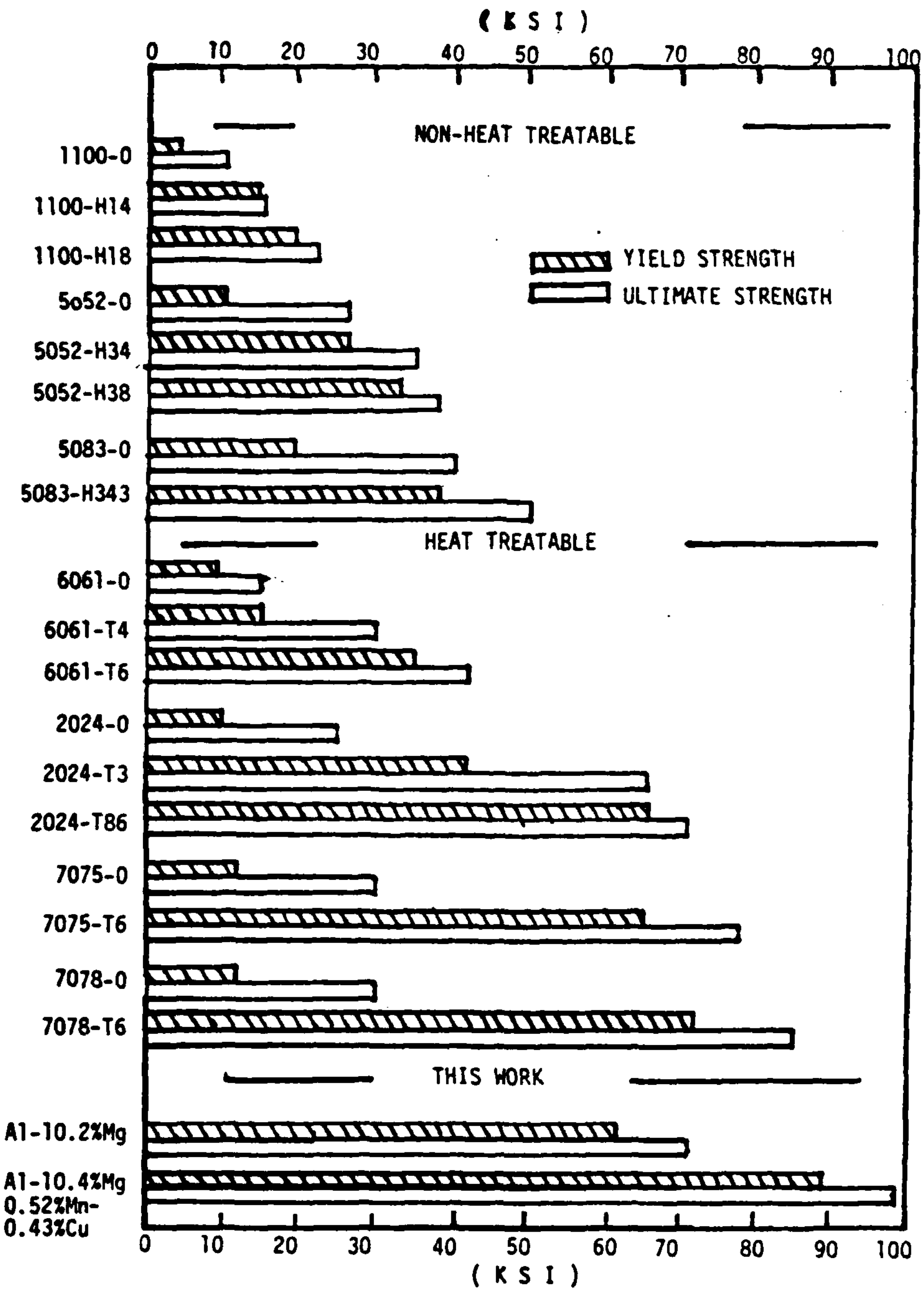


Fig. 14 Tensile and yield strengths for 063 sheet and comparative strengths from this work.

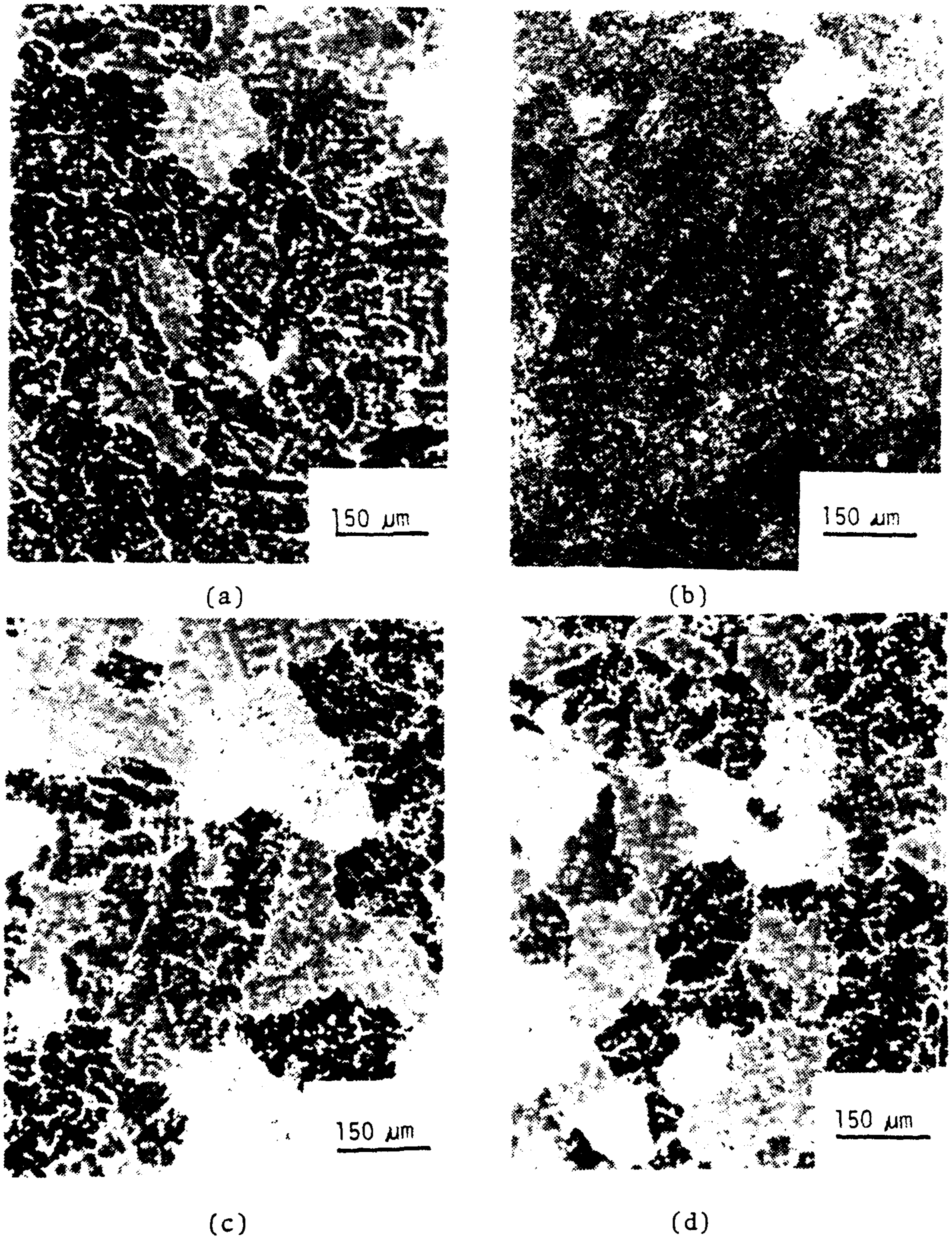
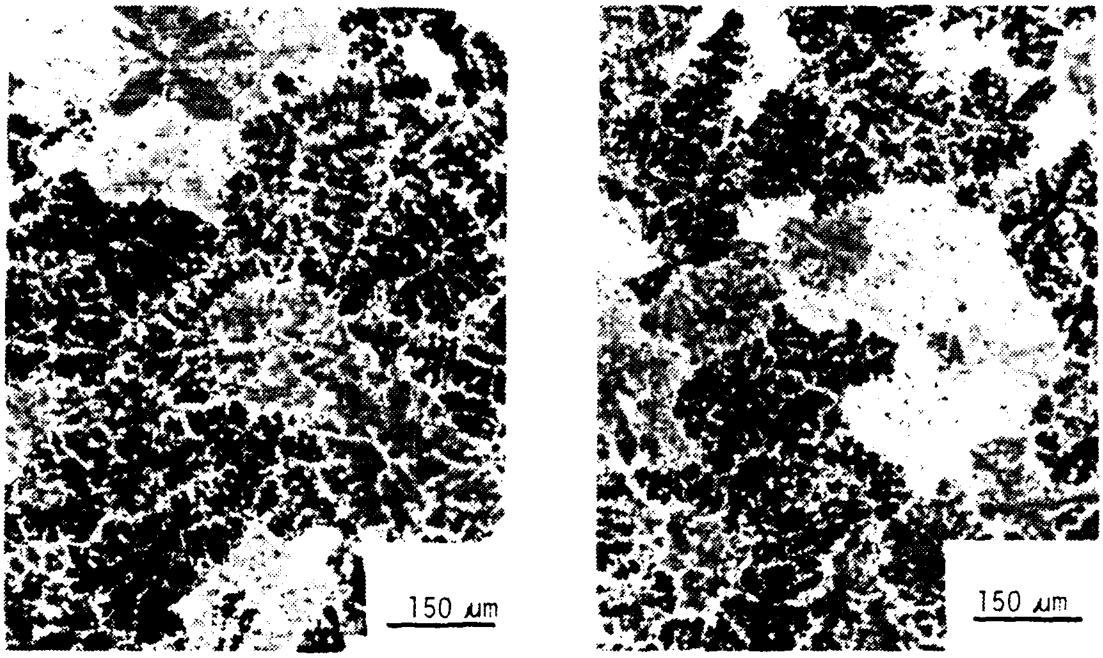


Fig. 15-1 As-cast microstructures of (a) Al-8.14%Mg-0.40%Cu, (b) Al-8.22%Mg-0.52%Mn-0.41%Cu, (c) Al-10.2%Mg, (d) Al-10.3%Mg-0.41%Cu.



(e) (f)

Fig. 15-2 As-cast microstructures of (e) Al-10.2%Mg-0.52%Mn, (f) Al-10.4%Mg-0.52%Mn-0.43%Cu.

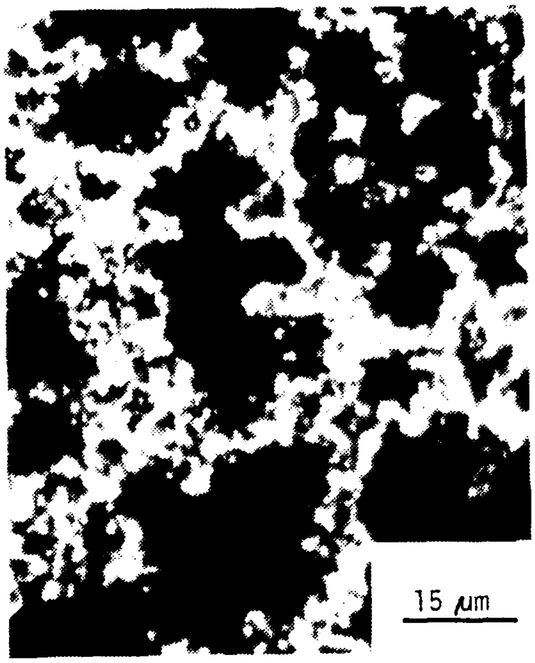


Fig. 16-a As-cast micro-
structure of Al-10.2%Mg.

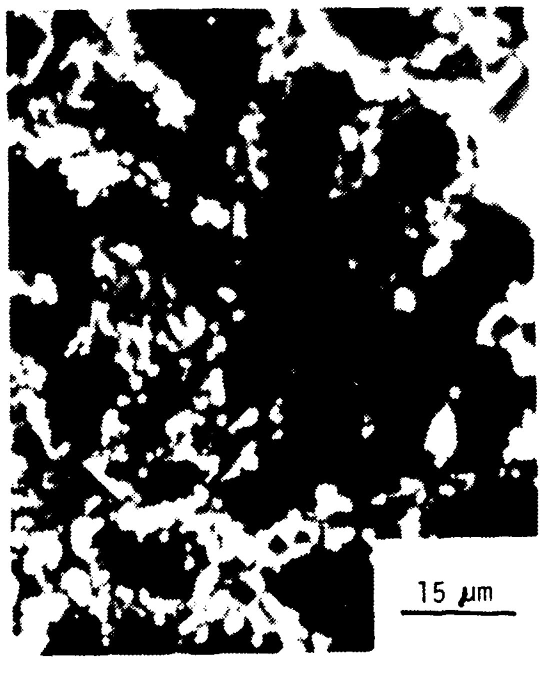


Fig. 16-b As-cast micro-
structure of Al-10.2%Mg.

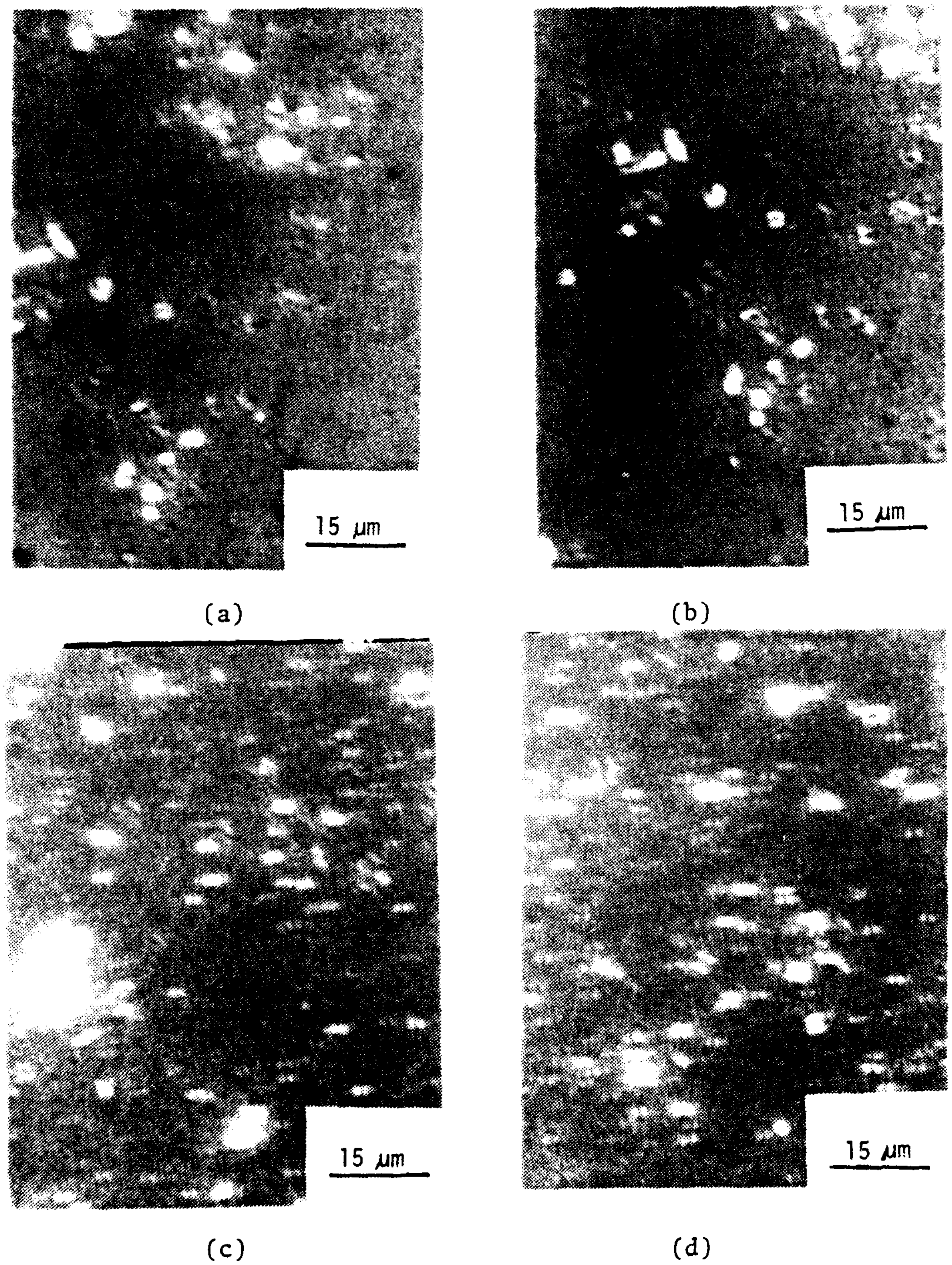


Fig. 17 Al-8.14%Mg-0.40%Cu after (a) and (b) oil quenching from 440°C after ten hours; (c) and (d) water quenching from 440°C after ten hours.

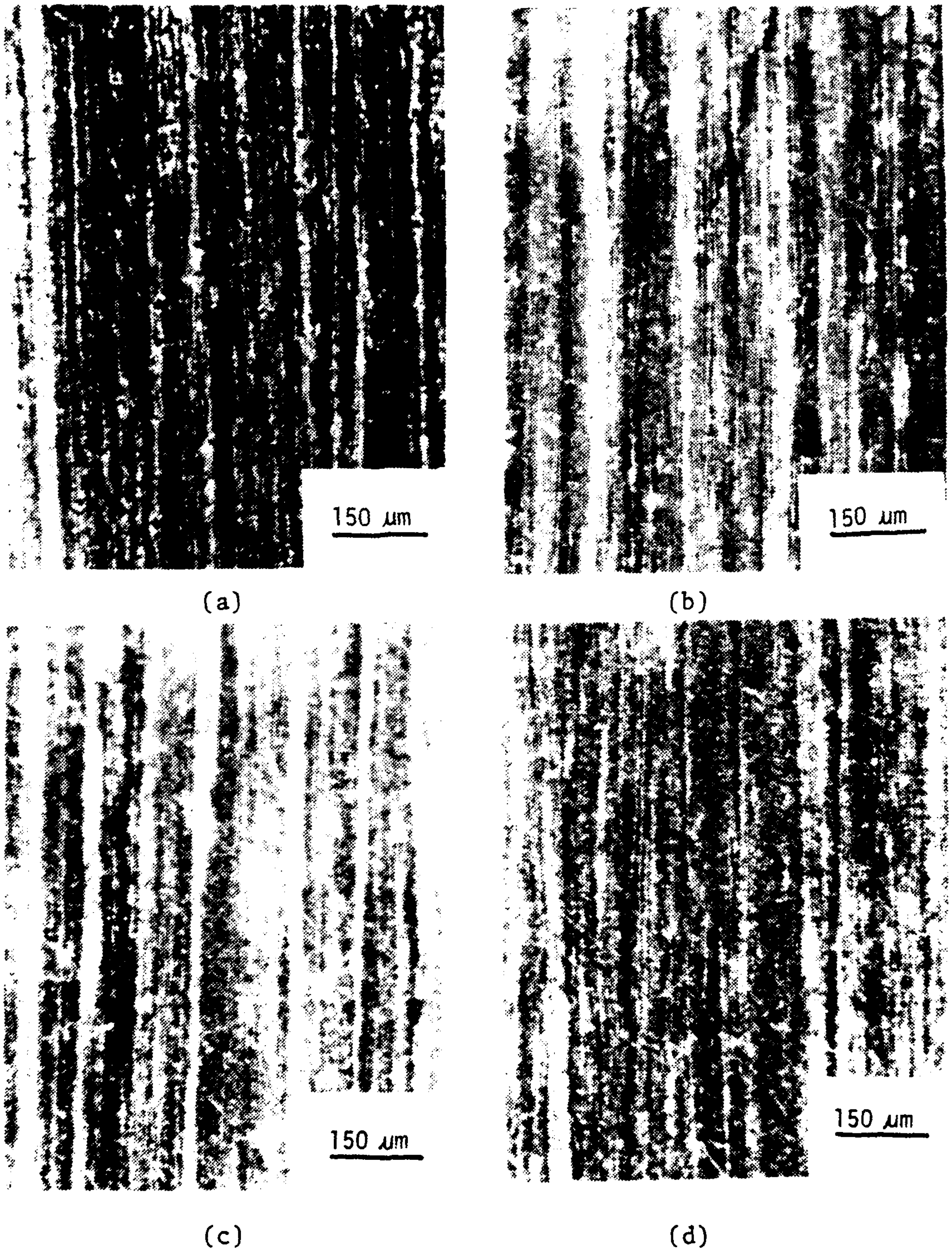
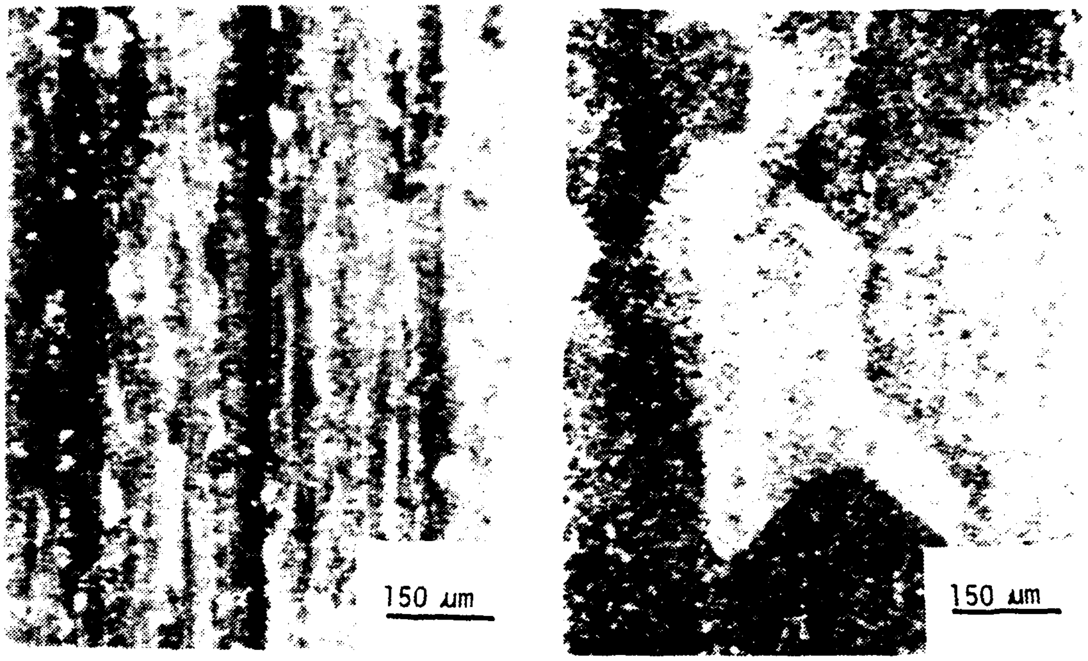


Fig. 18-1 Al-8.14%Mg-0.40%Cu warm rolled at (a) 200°C, (b) 250°C, (c) 300°C, and (d) 340°C using standard thermo-mechanical procedure.



(e) (f)

Fig. 18-2 Al-8.14%Mg-0.40%Cu warm rolled at (e) 380°C and (f) 420°C using standard thermo-mechanical process.

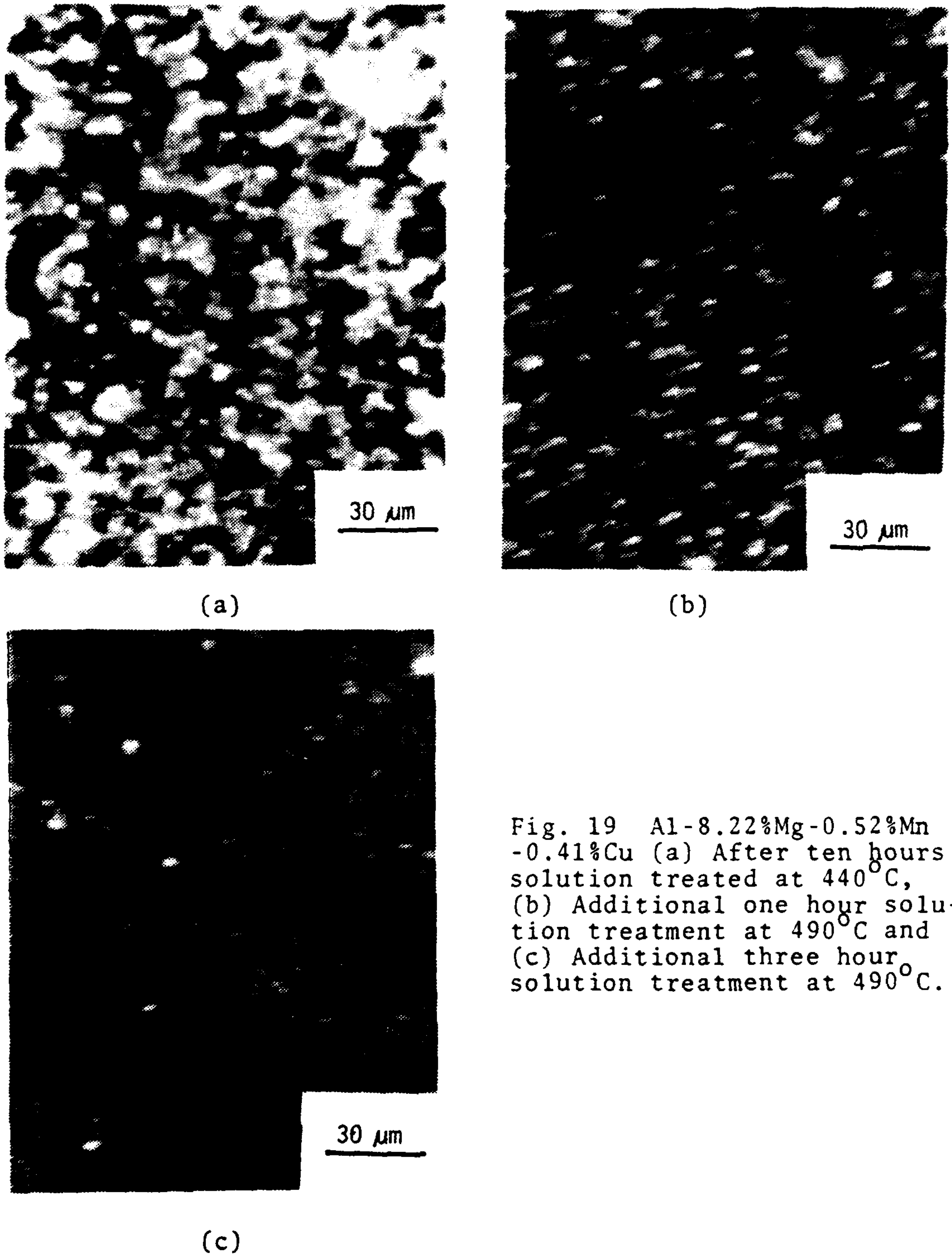


Fig. 19 Al-8.22%Mg-0.52%Mn -0.41%Cu (a) After ten hours solution treated at 440°C, (b) Additional one hour solution treatment at 490°C and (c) Additional three hour solution treatment at 490°C.

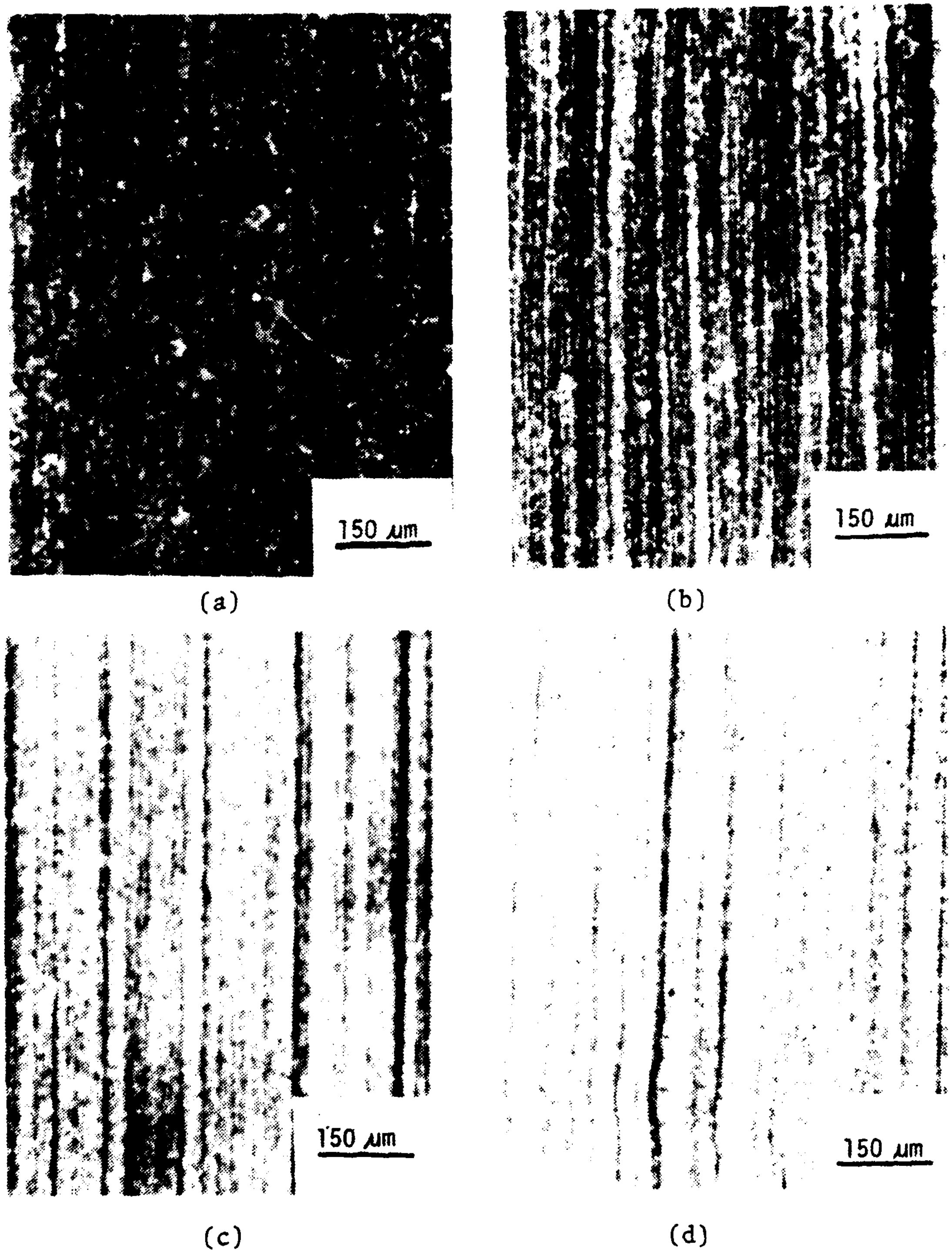


Fig. 20-1 Al-8.22%Mg-0.52%Mn-0.41%Cu warm rolled at (a) 200°C, (b) 250°C, (c) 300°C and (d) 340°C using standard thermo-mechanical process.

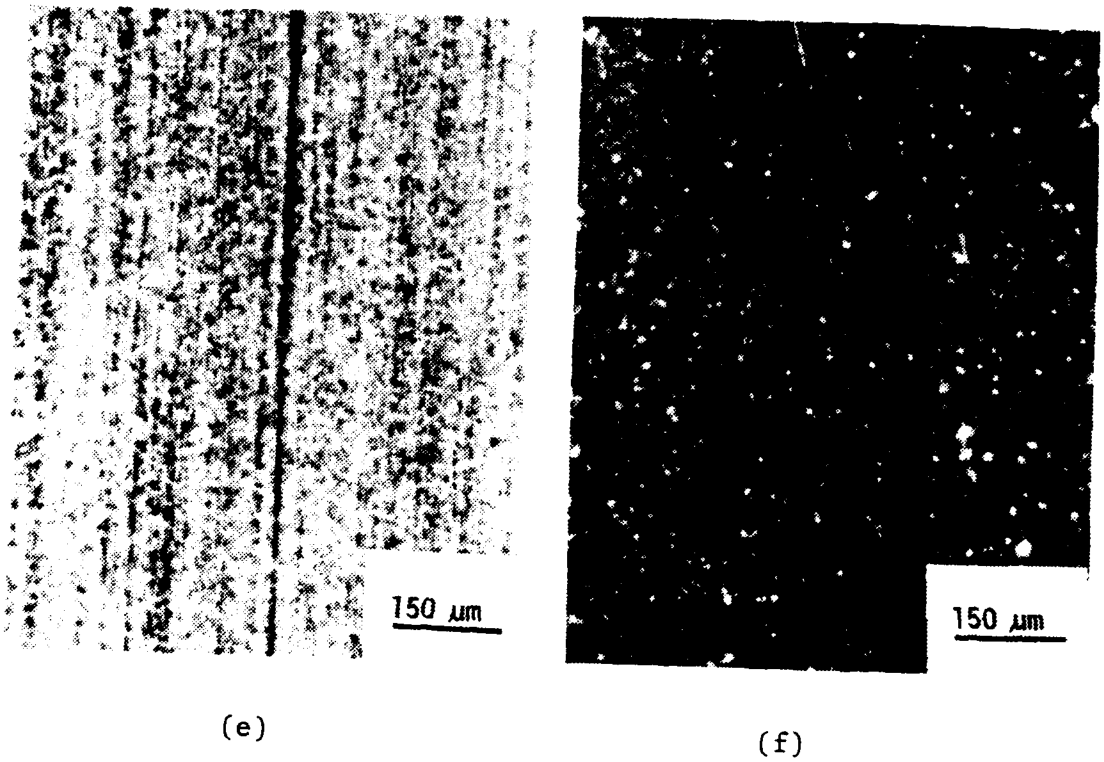


Fig. 20-2 Al-8.22%Mg-0.52%Mn-0.41%Cu warm rolled at (e) 380°C and (f) 420°C using standard thermo-mechanical process.

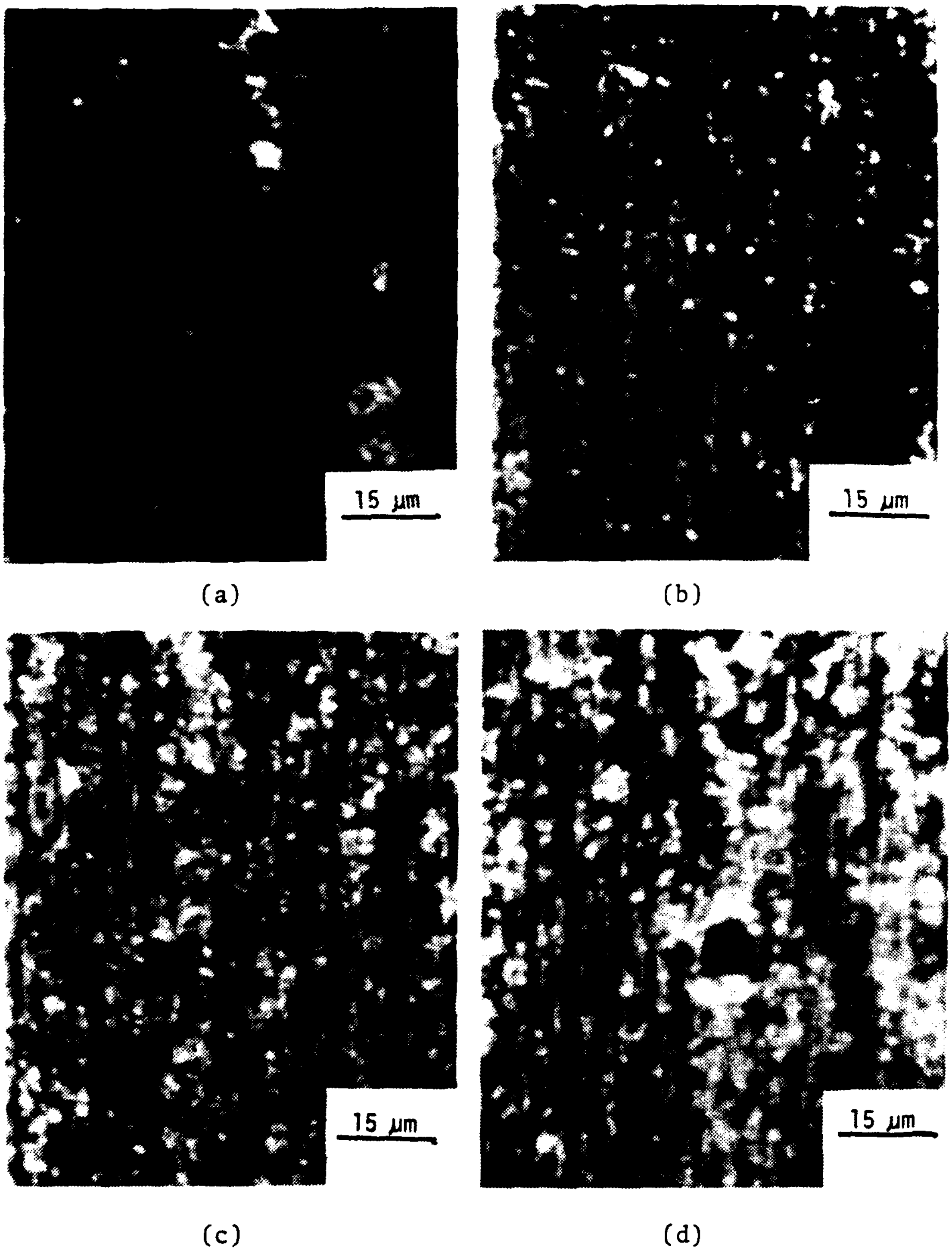
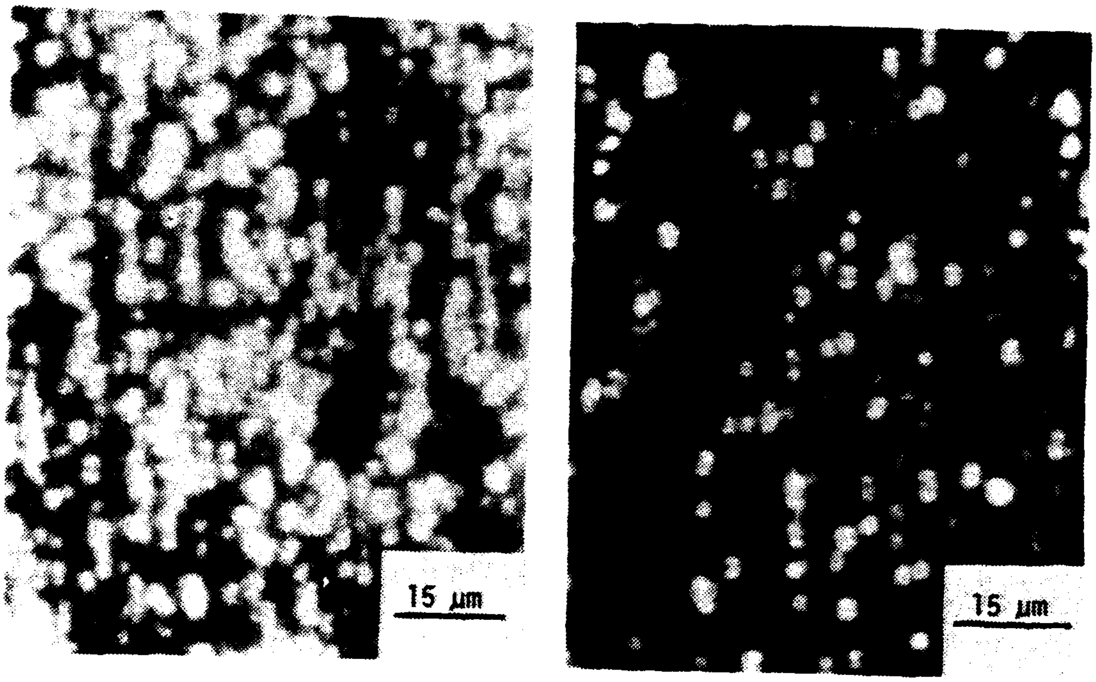


Fig. 21-1 Al-8.22%Mg-0.52%Mn-0.41%Cu warm rolled at (a) 200°C, (b) 250°C, (c) 300°C and (d) 340°C using standard thermo-mechanical process.



(e) (f)

Fig. 21-2 Al-8.22%Mg-0.52%Mn-0.41%Cu warm rolled at (e) 380°C and (f) 420°C.

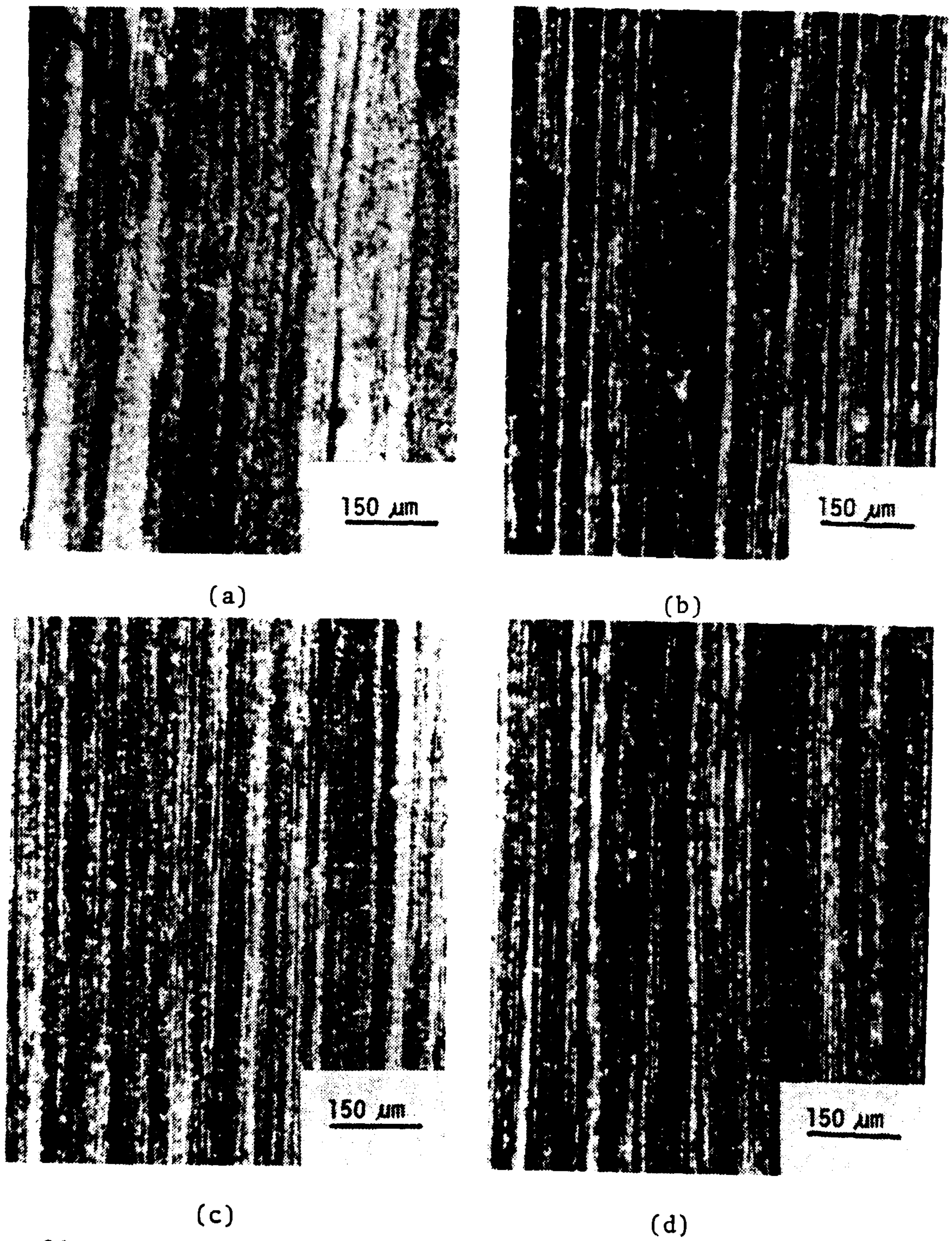
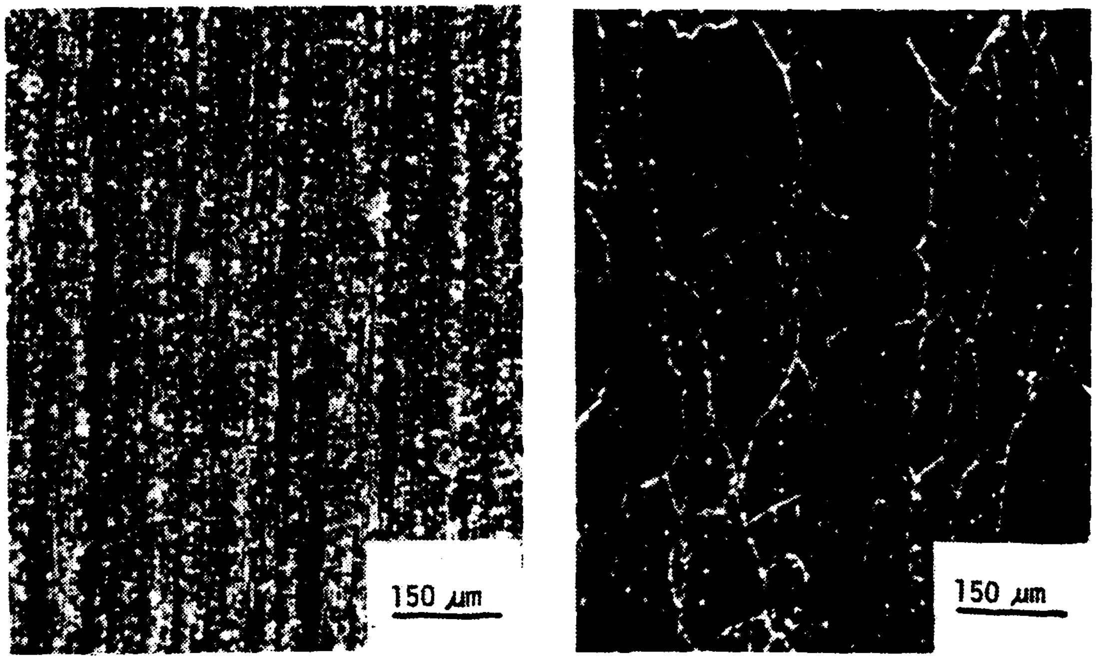


Fig. 22-1 Al-10.2%Mg warm rolled at (a) 200°C, (b) 250°C, (c) 300°C and (d) 340°C using standard thermo-mechanical process.



(e) (f)

Fig. 22-2 Al-10.2%Mg warm rolled at (e) 400°C and (f) 440°C using standard thermo-mechanical process.

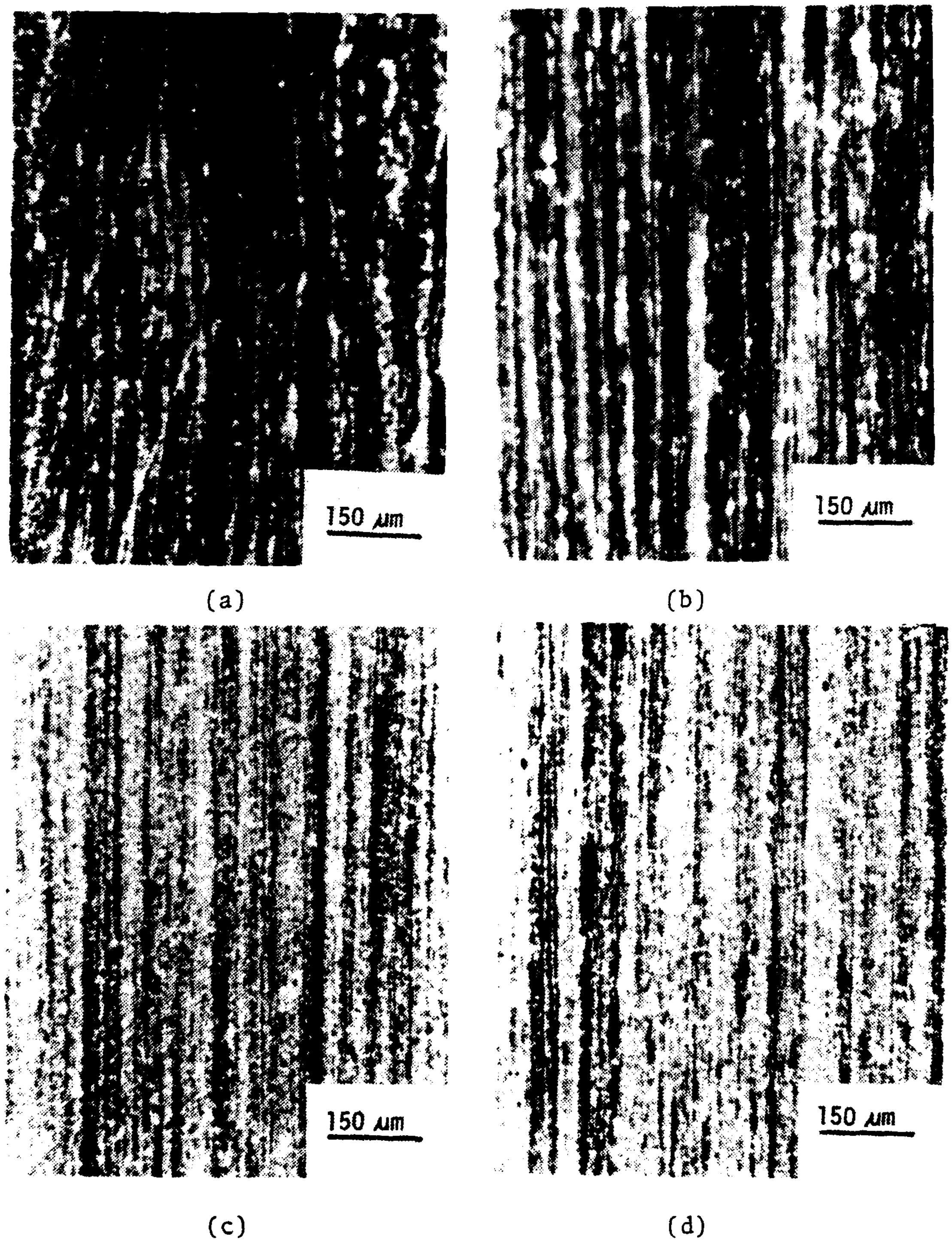
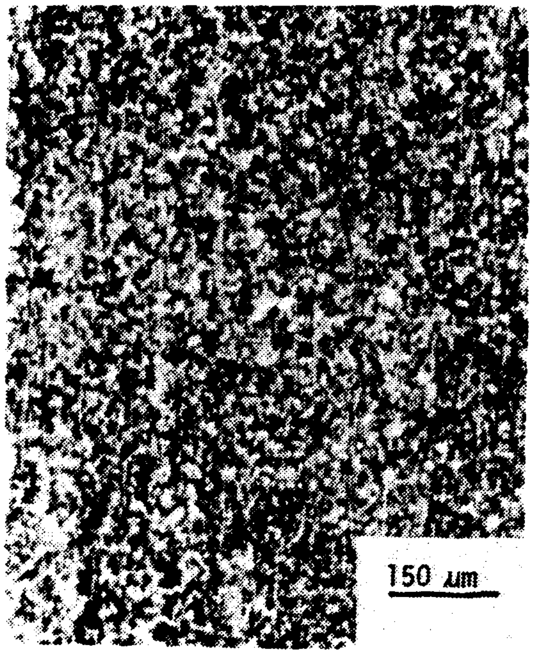
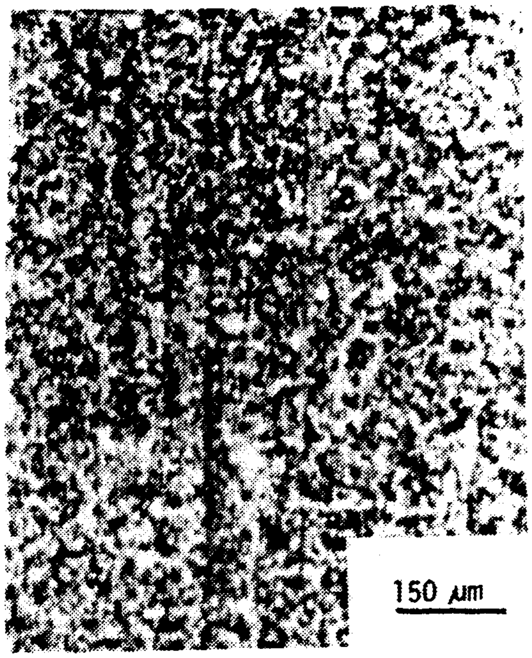


Fig. 23-1 Al-10.3%Mg-0.41%Cu warm rolled at (a) 200^{o}C, (b) 250^{o}C, (c) 300^{o}C and (d) 340^{o}C using standard thermo-mechanical process.

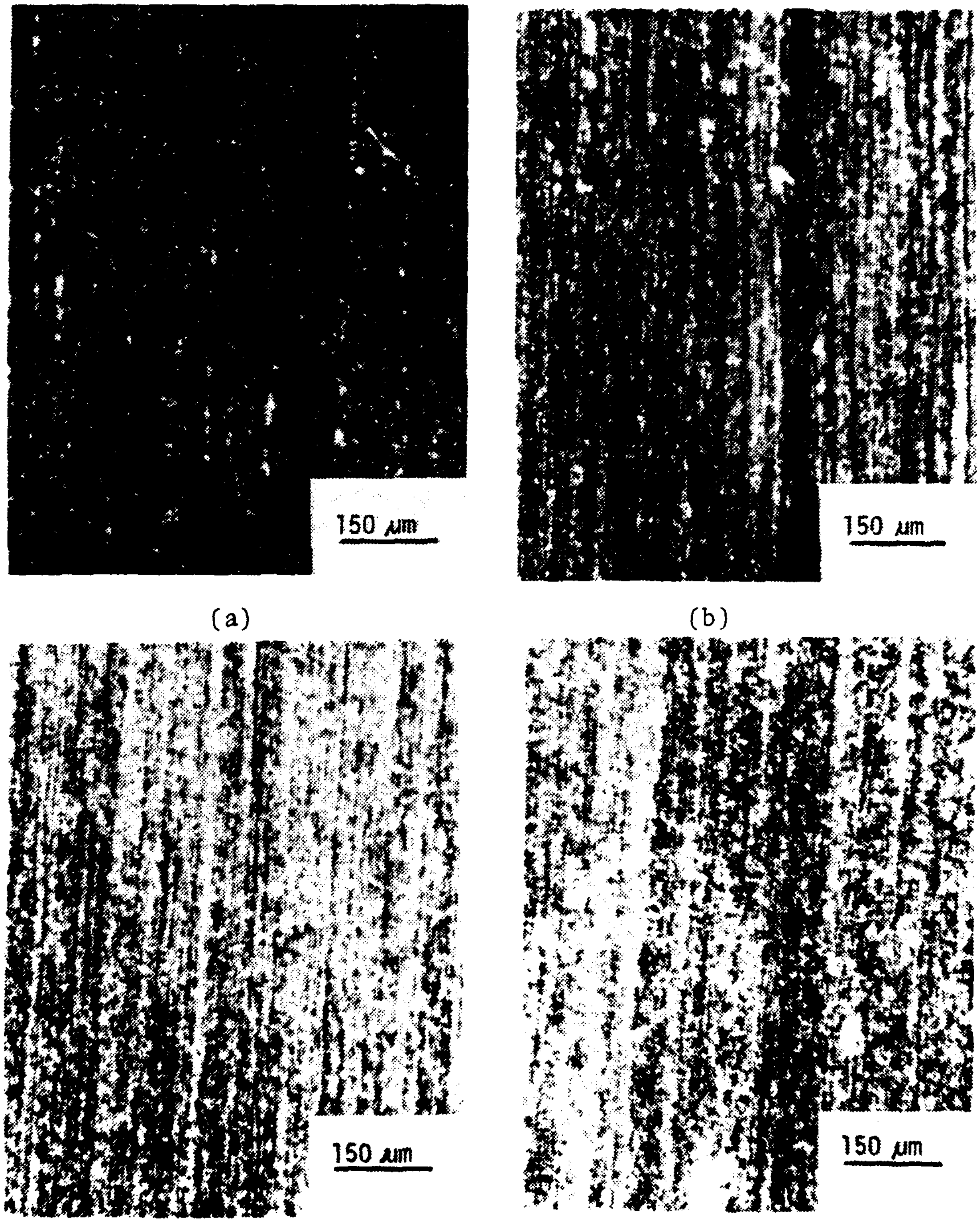


(e)



(f)

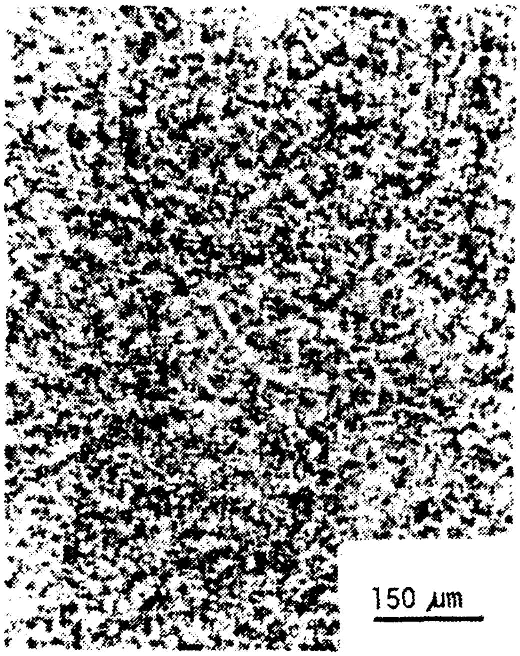
Fig. 23-2 Al-10.3%Mg-0.41%Cu warm rolled at (e) 400°C and (f) 440°C using standard thermo-mechanical process.



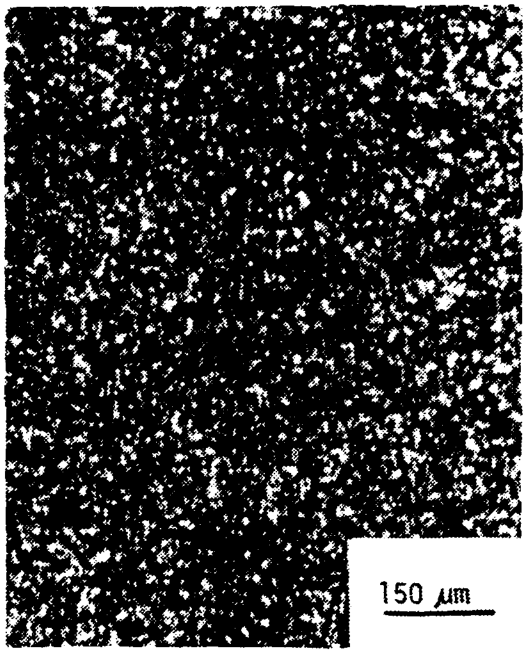
(a) (b)

(c) (d)

Fig. 24-1 Al-10.2%Mg-0.52%Mn warm rolled at (a) 200°C, (b) 250°C, (c) 300°C and (d) 340°C using standard thermo-mechanical process.

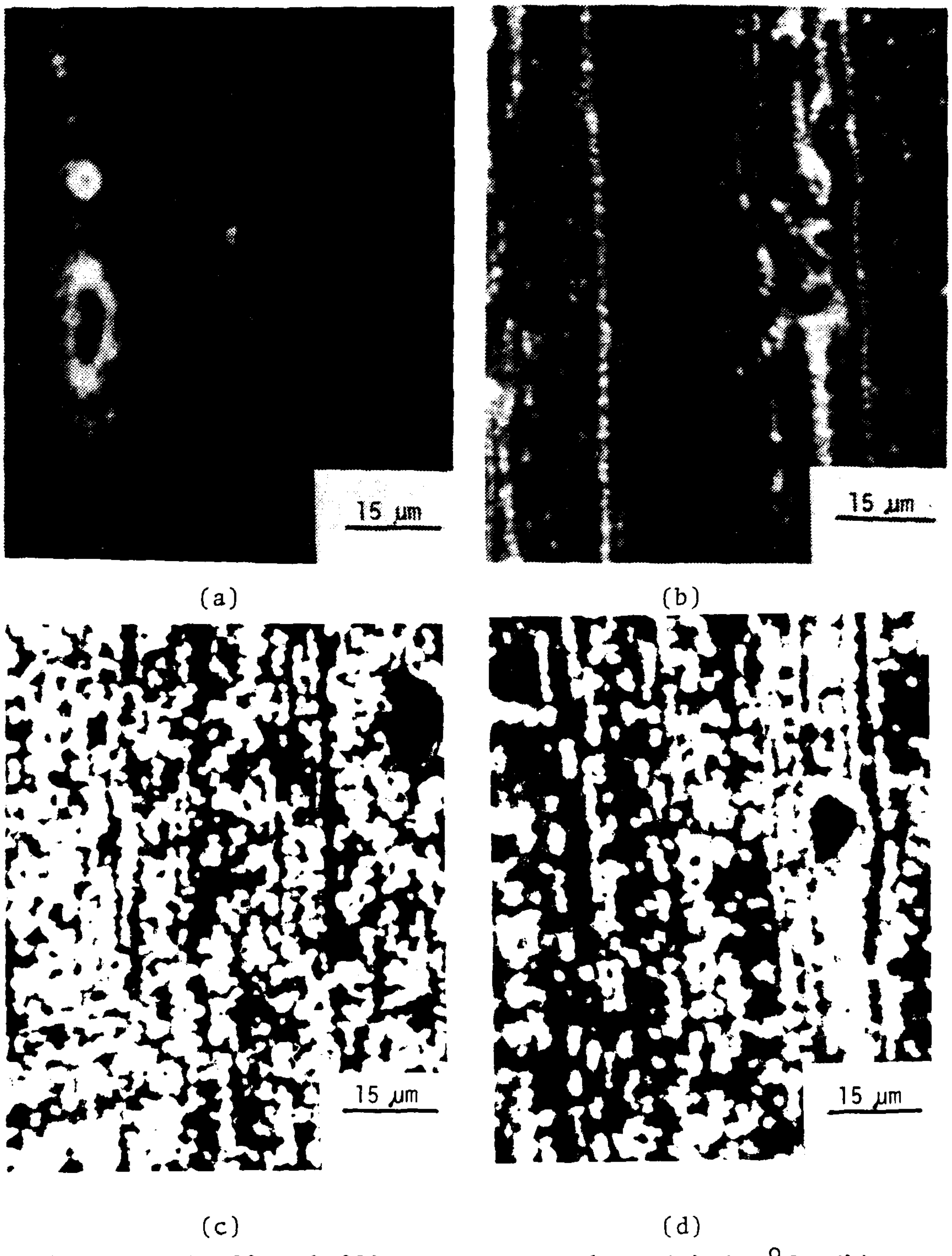


(e)



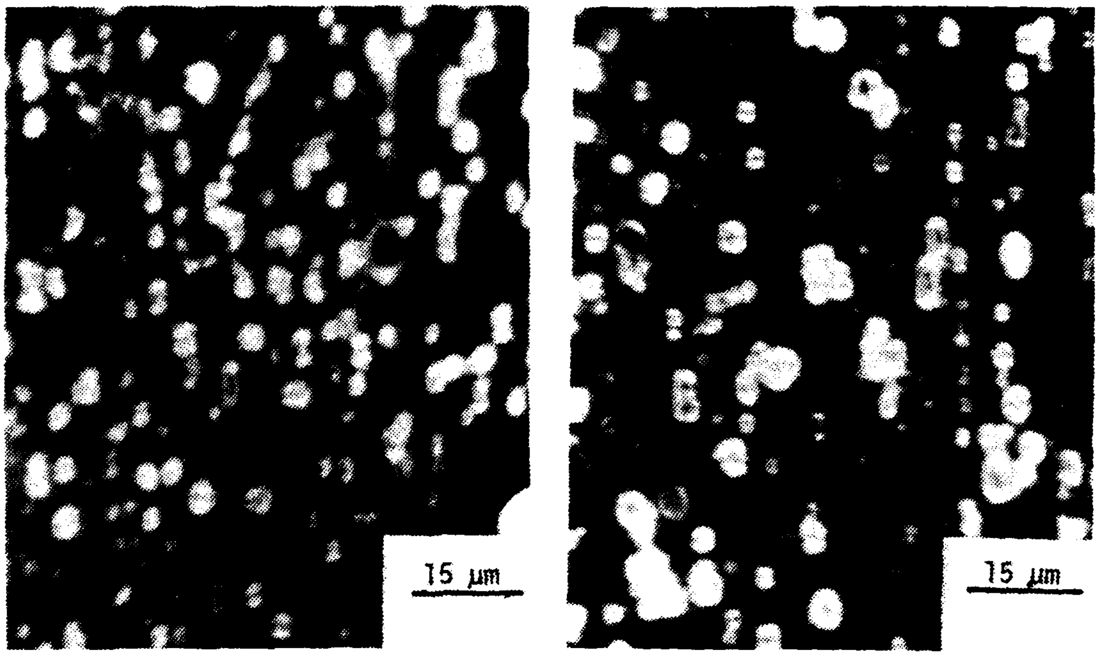
(f)

Fig. 24-2 Al-10.2%Mg-0.52%Mn warm rolled at (e) 400°C and (f) 440°C using standard thermo-mechanical process.



(a) (b) (c) (d)

Fig. 25-1 Al-10.2%Mg-0.52%Mn warm rolled at (a) 200°C, (b) 250°C, (c) 300°C and (d) 340°C using standard thermo-mechanical process.



(e) (f)

Fig. 25-2 Al-10.2%Mg-0.52%Mn warm rolled at (e) $400^{\circ}C$ and (f) $440^{\circ}C$ using standard thermo-mechanical process.

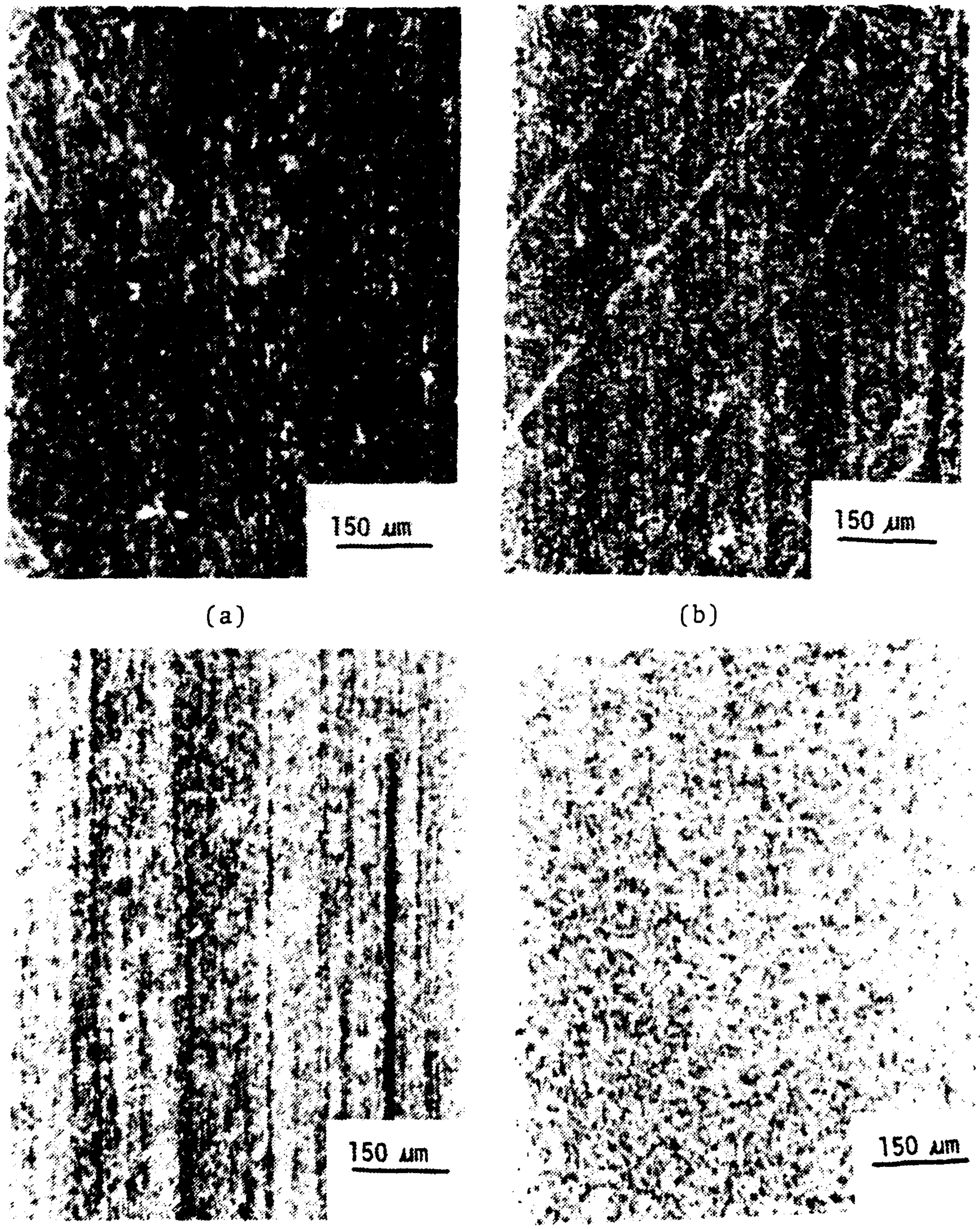
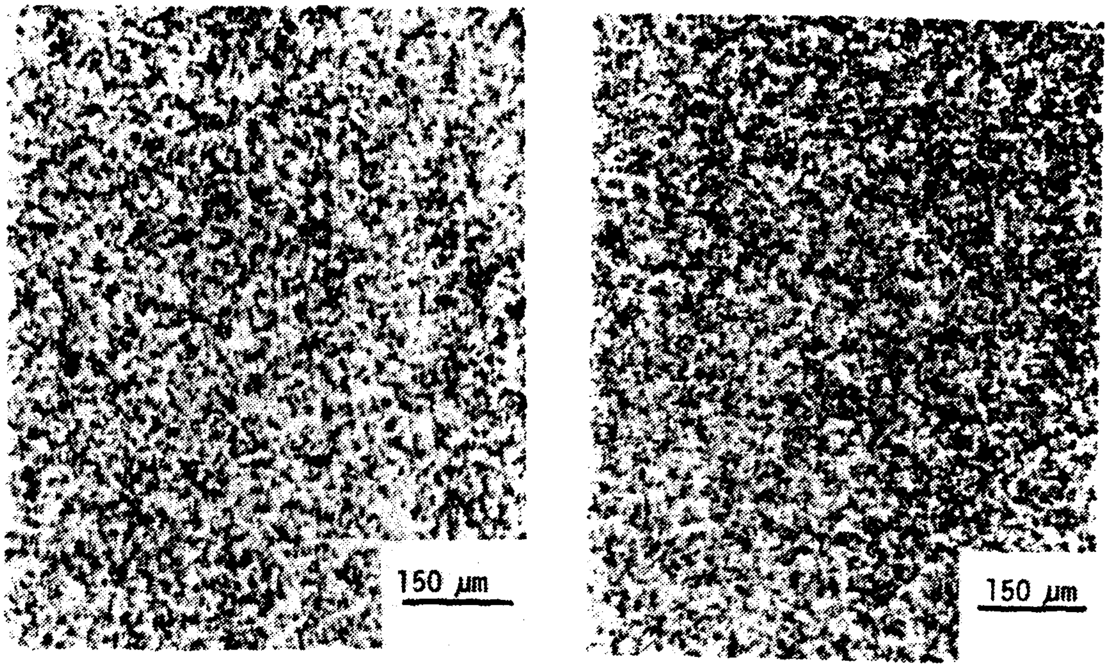


Fig. 26-1 Al-10.4%Mg-0.52%Mn-0.43%Cu warm rolled at (a) 200°C, (b) 250°C, (c) 300°C and 340°C using standard thermo-mechanical process.



(e) (f)

Fig. 26-2 Al-10.4%Mg-0.52%Mn-0.43%Cu warm rolled at (e) 400°C and (f) 440°C using standard thermomechanical process.

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